EAST ENGIN.
LIBRARY

TRANSACTIONS

of the American Society for Steel Treating

FEATURES

On the Double Diagram of the Iron-Carbon System —*Dr. Kotaro Honda* 183

Metals and Alloys for Industrial Applications Requiring Extreme Stability —*Jerome Strauss* 191

The Ac1 Range of Carbon Steel and Related Phenomena —*Axel Hultgren* 227

A Melting Record of Three Acid Open-Hearth Heats —*W. E. Griffiths and C. E. Meissner* 257

Application of Science to the Steel Industry—Section VI —*Dr. W. H. Hatfield* 278

Physical Chemistry of Rimmed Steel —*J. E. Carlin* 293

A Study of Burning and Overheating of Steel—Part I *W. E. Jominy* 298

Vol. XVI No. 2
August, 1929

TONCAN

Eliminate

maintenance and repair costs with

TONCAN *rust-resisting* IRON

MANY manufacturers have discovered to their great advantage the value of Toncan. The Steel Improvement and Forge Company, makers of automotive parts, is one. With an eye to the future, they accepted the advice of their contractor, J. L. Hunting Co., and decided upon Toncan for roofing material in the erection of their new plant.

Because Toncan—pure iron, scientifically alloyed with copper and molybdenum, is more highly resistant to rust and corrosion than any other ferrous metal.

There are hundreds of uses for Toncan. Roofing, gutters, cornices, window frames, metal lath, ventilating systems, etc. In or out of doors, wherever there is exposure to moisture or the elements · · · · you will note contractors have chosen Toncan.

Because they know its qualities, its endurance · · · its protection from fire and lightning. They know that after long years of service it will still retain its original neat appearance.

Let us mail you our booklet explaining Toncan Iron and its many uses. Write us.

CENTRAL ALLOY STEEL CORP. · *Massillon and Canton, Ohio*

World's Largest and Most Highly Specialized Alloy Steel Producers

REG. U.S. PAT. OFF.
TONCAN
COPPER
Molyb-den-um
IRON

The TRANSACTIONS *of the* AMERICAN SOCIETY FOR STEEL TREATING

Published and Copyrighted, 1929, by the AMERICAN SOCIETY FOR STEEL TREATING, *7016 Euclid Avenue, Cleveland, Ohio*

Issued monthly, $10.00 a year, $1.00 per copy

Entered as second-class matter, February 7, 1921, at the post-office at Cleveland, Ohio, under the Act of March 3, 1879

RAY T. BAYLESS, *Editor*

Vol. XVI August, 1929 No. 2

Table of Contents

On the Double Diagram of the Iron-Carbon System—By Dr. Kotaro Honda.. 183

Metals and Alloys for Industrial Applications Requiring Extreme Stability —By Jerome Strauss 191

The Ac1 Range of Carbon Steel and Related Phenomena—By Axel Hultgren 227

A Melting Record of Three Acid Open-Hearth Heats— By W. E. Griffiths and C. E. Meissner 257

The Application of Science to the Steel Industry—Section VI— By Dr. W. H. Hatfield 278

The Physical Chemistry of Rimmed Steel—By J. E. Carlin..... 293

A Study of Burning and Overheating of Steel—Part I—By W. E. Jominy... 298

Recommended Practice Committee Releases 314

Reviews of Recent Patents 328

Engineering Index 331

News of the Society .....33 Adv. Sec.

News of the Chapters .....36 Adv. Sec.

Founder and Honorary Members .....19 Adv. Sec.

Henry Marion Howe Medalists .....19 Adv. Sec.

National Officers and Board of Directors .....19 Adv. Sec.

Sustaining Members .....20 Adv. Sec.

Standing Committees .....28 Adv. Sec.

Chapters and Officers .....30 Adv. Sec.

Items of Interest .....40 Adv. Sec.

Employment Service Bureau .....43 Adv. Sec.

Clearing House .....43 Adv. Sec.

Index to Advertisements .....95 Adv. Sec.

"Yes Sir, I got that idea at last year's National Metal Congress . . . it saved us $10,000 this year"

INSTITUTE of METALS DIVISION of the A. I. M. E.

IRON and STEEL DIVISION of the A. I. M. E.

AMERICAN WELDING SOCIETY

IRON and STEEL DIVISION of the A. S. M. E.

AMERICAN SOCIETY for STEEL TREATING

THAT'S just one company's experience. Hundreds of other visitors to last year's National Metal Congress carried back ideas to their plants that are now increasing profits through time, labor and material savings.

Every late development in the production, fabrication and use of metals will be discussed during the technical meetings of the five leading societies of the metal industry, who are holding simultaneous meetings during this year's

NATIONAL METAL CONGRESS

During the same week the greatest annual exposition of the metal industries will be held during the week of *September 9 in Public Auditorium, Cleveland.* Over 275 manufacturers will exhibit the latest machinery, equipment and methods used in metal working and manufacturing industries. That's why no manufacturer who works with or makes *any kind of metal in any form* can afford to miss the eleventh annual National Metal Congress and

NATIONAL METAL EXPOSITION

NATIONAL METAL CONGRESS has been planned to give to American Industry, at one time and in one place, all the latest developments and progress of the metal industries. Instead of five or more meetings, that would take five or more trips and weeks, one trip and one week's time covers everything. In no other way, at so small a cost, can you and your men obtain this valuable information—the hundreds of profit making ideas—that the National Metal Congress will give you.

CLEVELAND—WEEK OF SEPTEMBER 9

Plant Inspection Trips

Many of Cleveland's leading industrial plants will be open to visitors. Competent guides will be available to conduct these inspection trips, to explain in detail the various methods and processes in operation. A complete list of these trips will be given in the Exposition program, a copy of which will be given you at the entrance to the National Metal Exposition.

TRANSACTIONS
American Society for Steel Treating

VOL. XVI AUGUST, 1929 NO. 2

ON THE DOUBLE DIAGRAM OF THE IRON-CARBON SYSTEM

By Dr. Kotaro Honda

Abstract

The double diagram of the iron-carbon system is widely used in Europe and America, especially in Germany, where it was born and brought up by noted metallurgists. From the experimental point of view, however, the direct precipitation of graphite from melt is very doubtful; moreover, the result of the X-ray analysis does not accord with the consequences of the double diagram. From these and other facts, the present author concluded that for pure iron-carbon system, the single diagram is the simplest and the most satisfactory one.

IT is a curious fact that notwithstanding a great advancement of the science of iron and steel in recent years, a very important and fundamental question regarding the equilibrium diagram of the pure iron-carbon system still remains unsettled.

When cast iron is cooled from melt, carbon is present partly as graphite and partly as cementite; the quantity of graphite is larger as the velocity of cooling diminishes, and the quantity of silicon present in the cast iron increases. In regard to the formation of graphite we have two controversial theories; in one theory, graphite is considered as a product directly precipitated from the melt, and in the other the first precipitation product is assumed to be the cementite, the graphite being its decomposition

A paper presented before the Western Metal Congress, Los Angeles, California, January 14 to 18, 1929. The author, Dr. Kotaro Honda, honorary member of the society, is professor of metallurgy, Tohoku Imperial University, Sendai, Japan. Manuscript received January 15, 1929.

product at a high temperature. In the first theory, a double diagram corresponding to the stable and the metastable equilibrium is adopted, while in the second theory, we have a single diagram. In Fig. 1, the full lines correspond to the single diagram belonging to cementite, while the dotted lines are those belonging to the graphite; these two systems of lines forming the double diagram. Since the cementite decomposes at high temperatures, graphite lines are said to form a stable diagram, and the cementite lines to form a metastable one.

The double diagram of the iron-carbon system has been developed by Charpy, Grenet, Heyn and others, and is based on the following facts: (1) Since cementite decomposes into iron and graphite at high temperatures, the diagram of the system iron-cementite is not a stable one. (2) In the case of hypereutectic cast iron, the form of the graphite is straight needle-shaped and just what is expected from the form of a primary crystal. (3) In a cast iron containing silicon of the amount which is a little less than the eutectic concentration, the form and the distribution of graphite are very similar to an eutectic form. (4) Ruer and Goerens found in their thermal analysis for cast iron of eutectic composition that by repeated heating and cooling through the solidifying range, two stepped halting points appear in the heating curve; they held the higher point as the graphite line and the lower one as the cementite line. (5) In the thermal analysis of iron-carbon system, it has often been found that a small thermal effect is observable at 30 to 40 degrees Cent. above the A_1 line. Ruer considers this to be the eutectoid line of graphite.

For a period of more than ten years, the present writer has occasionally published his view against the double diagram of the iron-carbon system; the results of recent investigations have further led him to confirm the above view, as shown below.

(1) Many compounds decompose at high temperatures into a simpler form; that is, they have a degree of stability. Cementite has also a certain degree of stability. Hence if the first precipitation product be always the cementite, but not the graphite, the single diagram must be adopted in spite of the decomposition of cementite at high temperatures.

(2) The question whether the decomposition of cementite is solely due to temperature or to the catalytic action of a gas, such

as CO or CO_2 at the high temperatures, is not yet definitely solved. According to the result of investigation by Prof. T. Murakami and the present writer,[1] it seems very probable that the decomposition of cementite is not solely due to temperature such that

$$Fe_3C \rightarrow 3Fe + C,$$

but it is caused by the catalytic action of CO_2 gas such that

$$Fe_3C + CO_2 \rightarrow 3Fe + 2CO \rightarrow 3Fe + C + CO_2$$

Thus, if a small quantity of CO_2 gas be initially present within cast iron, it reacts catalytically and the graphitization goes on

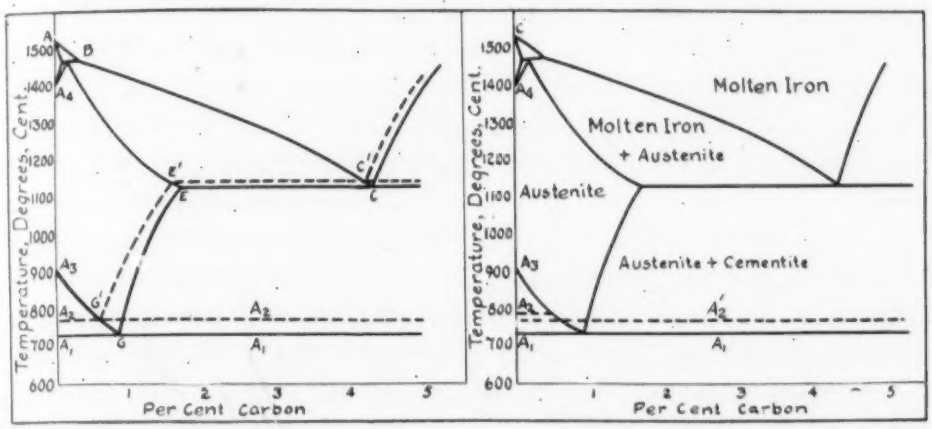


Fig. 1 Fig. 2

Fig. 1—Double Iron-Carbon Diagram as Developed by Charpy, Grenet, Heyn and others. Fig. 2—Single Iron-Carbon Diagram as Proposed by the Author.

continuously. At any rate, the results of several investigators agree in the fact that the graphitization is accelerated by the presence of CO or CO_2 gas. If the decomposition of cementite is really due to the catalytic action of these gases, the single equilibrium diagram must be the correct one.

(3) As we have explained above, the straight graphite observable in hypereutectic cast iron has a form characteristic to a primary crystal, but it must be remarked that this graphite is not a single crystal, but that it consists of an aggregate of fine particles of graphite, the orientation of which is directed uniformly in all directions.[2] This fact favors the view that the graphite is a decomposition product of cementite. The formation of the

[1]Science Reports of the Tohoku Imperial University, Sendai, Japan. Vol. X, 1921, p. 273.

[2]Weaver, Mitteilungen, Kaiser-Wilhelm Institut.

straight graphite is explained in the following way: Just after the solidification of a hypereutectic cast iron, a certain number of primary straight cementite crystals are found in parallel, the interspace of which is filled up by eutectics. At high temperatures the cementite in eutectic is first decomposed into austenite and graphite particles, and then the primary cementite decomposes from its boundary. By virtue of surface tension between the graphite and the austenite, these graphite particles assemble themselves by lateral movements into a straight graphite. This view was directly confirmed by an experiment made by the present author,[3] in which by a rapid cooling from the melt, white cast iron was first made, and then annealed at 1000 to 1100 degrees Cent. (1832 to 2012 degrees Fahr.) for several intervals of time, at the end of which the specimens were quenched in cold water. The microstructure of these specimens showed different grades of the formation of the straight graphite and confirmed the above view.

(4) When a hypoeutectic cast iron containing a suitable amount of silicon was gradually cooled from the melt, fine graphite flakes in the form of whirls were found. This structure is usually considered as an eutectic of graphite and austenite, and is called the eutectic graphite. But according to an investigation of Prof. M. Hamazumi,[4] this eutectic is not an eutectic between graphite and austenite, but a decomposition product of eutectic between cementite and austenite. The cementite in the eutectic is just after the solidification, decomposed into austenite and graphite; these graphite particles assemble on the grain boundary of austenite by virtue of surface tension and thus form graphite in whirls.

(5) As we have remarked above, there is a small heat change at 30 to 40 degrees Cent. above the A_1 line, which is independent of carbon content. This point is not the graphite eutectoid as Ruer considers it to be, but a point, which the present author has called the A'_2 point,[5] and where the paramagnetism of austenite abruptly increases. Since this abrupt change of paramagnetism is a consequence of some internal change, it must be accompanied by a thermal change.

[3] Not yet published.

[4] Science Reports of the Tohoku Imperial University, Sendai, Japan. Vol. XIII, 1924, p. 133.

[5] Science Reports of the Tohoku Imperial University, Sendai, Japan. Vol. XV, 1926.

(6) Recently Dr. T. Kase and the present author[6] repeated Ruer-Goerens' experiment, using the same method and the same material; but we could not generally confirm their results. In a few cases, however, we observed a two-stepped heat absorption in the heating curve, but never in the cooling curve. The differences of temperature in the two steps amounted to only 3 to 5 degrees. This phenomenon is explained by the difference of the velocity of dissolution of cementite and graphite in iron. During heating of partially graphitized cast iron, cementite begins first to form an eutectic with iron near the eutectic point and then graphite begins to do so, and thus a two-stepped heat change is observed. But since this difference of temperatures is caused by that of the velocity of dissolution of these substances, it diminishes with the slowness of heating and thus vanishes in the case of an extremely slow heating. Hence this stepped thermal change cannot be considered as the eutectic points of cementite and graphite respectively.

(7) According to the result of X-ray analysis, when cementite or graphite is dissolved in iron, forming austenite, carbon atoms do not replace iron atoms in its space lattice by a simple substitution, but are present in the interspace of the lattice, that is, occupy the center of the face-centered cubic, scattered here and there according to the law of probability. Thus, whether cementite or graphite dissolves in iron, the result is the same austenite, and hence there must be one and only one temperature at which, during cooling, carbon atoms precipitate as cementite, so that two distinct lines corresponding to cementite and graphite solubilities cannot theoretically be admissible. That is, the existence of two lines, GE and G′E′ in Fig. 1 is not allowable; in fact, they can never be found experimentally. If there is only a single cementite solubility line, the graphite lines in the double diagram cannot be completed.

(8) As regards the double diagram, the question of whether in molten cast iron, carbon is present as carbon atoms or cementite molecules, is very important. If a substance has a transformation point at a high temperature, the stable phase above this temperature can be obtained at room temperature by quenching, either in the pure phase or in the one mixed with the transformed phase. In-

[6]Not yet published.

versely, when a substance is quenched from a high temperature and gives rise to another phase at room temperature, it is to be concluded that this new phase is the one which is stable at the higher temperature. Now, if a molten cast iron is rapidly cooled by pouring it into a thin metallic mold, cementite is always found, but not graphite, hence it may be concluded that in molten cast iron, carbon is present as cementite molecules, Fe_3C, but not as carbon atoms. The interesting fact discovered by Y. Matsuyama[7] that an alloy which, in solid phase, exists as an intermetallic compound, is always found as such in molten state, confirms also the above view.

The above conclusion contradicts obviously to the consequence of the theory of the double diagram, in which it is necessary to assume that in molten state, at least a part of the carbon is present in the atomic state. Otherwise, we must assume that just before the solidification, a part of the cementite is to be decomposed; this is a very improbable assumption.

(9) According to an investigation by Prof. T. Murakami and the present author,[8] a hypoeutectic cast iron, when quenched in water from a temperature, at which the solidification is just completed, that is, from 1130 degrees Cent. (2066 degrees Fahr.), contains almost no graphite, but when it is further cooled slowly to 1050 to 1000 degrees Cent. (1922 to 1832 degrees Fahr.), and then quenched, the graphitization is very marked. If the graphite is to be precipitated from the melt, even the specimen quenched from 1130 degrees Cent. should show the same degree of graphitization, but this is not actually the case.

From these facts, so far enumerated, we may safely conclude that the adoption of the double diagram is not only unnecessary, but unreasonable. Especially in the field below 1.7 per cent carbon, there are no experimental facts confirming this diagram. Hence, we must adopt the single diagram as the simplest and satisfactory one. Fig. 2 is the most reliable iron-carbon diagram based on the recent data.

In this figure, the following remark deserves a notice. In the iron side, two horizontal broken lines are drawn at 790 and 770 degrees Cent., respectively; they are not graphite lines, but the former is the A_2, or the critical point of pure iron, and the

[7]Science Reports of the Tohoku Imperial University, Sendai, Japan. Vol. XVI, 1927.

[8]Science Reports of the Tohoku Imperial University, Sendai, Japan. loc. cit.

latter the A_2' point of austenite, at which the paramagnetism of austenite begins to increase rapidly. Since the field A_3GA_1 consists of a mixture of pure iron and the austenite of a definite concentration, these two lines must be horizontal. The austenite of different concentrations, which lie between the A_3 and solubility lines, has an almost constant value for A_2' point, and hence the lower line is drawn horizontal. As the amount of pure iron in the triangular field rapidly diminishes as the carbon content increases, the A_2 line becomes, in actual case, soon inconspicuous, the result being as if the A_2 line falls at first so the A'_2 level and afterwards remains constant. The third broken line at 215 degrees Cent. shows the critical point of cementite, and is, therefore, horizontal. These three lines do not show any change of phase, a change going on in the atoms themselves, this change being a definite function of temperature, and hence they cannot be governed by the phase rule. To distinguish them from lines belonging to the change of phase, they are drawn in broken lines. Other lines in the diagram do not require any explanation, as they are well-known.

DISCUSSION

Discussion: By Dr. Zay Jeffries, president of the American Society for Steel Treating.

Mr. Chairman, ladies and gentlemen, I know as we listened to Dr. Honda make his presentation many of us may have been embarrassed by lack of understanding of the terms themselves, and perhaps somewhat more embarrassed by lack of understanding of Dr. Honda himself, but imagine how much better he has been able to do in our language then we could have done in Japan in his language. Now the paper which Dr. Honda has presented has rather important consequences in connection with the simplicity of presentation of the involved subject of the equilibrium diagram of iron and carbon. It is the most important constitution diagram of all the metallic systems, and more work has been done on it perhaps than on all other systems combined. The old double diagram has gone through many changes and has taken many specific forms, and the fact that there has been so much disagreement on the significance of the various lines and the positions of the various points is proof that it is not simple. Coupled with that, we have the proposition that the double diagram is never associated with the binary system of iron-carbon. It is always associated with some ternary or more complex system. Why then should we necessarily be wedded to a double diagram of the iron-carbon system when we must have a third element present to develop the double diagram? The diagram presented by Dr. Honda as the single diagram certainly is the type of diagram which adequately takes care of the binary iron-carbon system. Dr. Honda has brought before us today, in addition to a general plan for simplification, the presentation of a few new facts which lead him to the

conclusion that we can better afford to adopt the single diagram as the working diagram. He has made some assumptions which I think are unnecessary and which may detract somewhat from the acceptance of the idea of the single diagram for a good many purposes. The facts that he brought out are briefly: 1. That the A_2 point can be explained on the basis of a magnetic change. 2. That in the history of solidification it is the cementite which first forms, and the graphite then forms by decomposition of the cementite. The evidence he has given on that point is quite strong, that is, when the cast iron is quenched immediately after solidification, he finds practically no graphite, but much cementite. On reheating to a high temperature, but below the melting point, graphite forms in the exact location of the cementite. This indicates that the graphite is a secondary formation. Now in view of the fact that cementite is the first form of carbon to precipitate, Dr. Honda feels more or less obligated I think to propose that the cementite must be present as molecules in the liquid. He also gives some evidence of his own and his associates to support the view that any metallic alloy system having an intermetallic compound as one of the components or constituents, retains the elements in molecular form when the alloy is melted. I think this conclusion itself is subject to quite serious objections and need not be considered as a necessary part of the adoption of the single diagram in the iron-carbon binary system. To elaborate on that point just a little, I would like to call your attention to the two fundamentals of compound formation. One is combination in accordance with stability, and the other is combination in accordance with availability of the combining atoms. Oxygen combines with iron to form rust. That does not mean that iron oxide is the most stable compound of oxygen. Supposing some aluminum were present in the neighborhood of the particular atom or molecule of oxygen which combined with the iron, the aluminum oxide may have formed, and if so it would be a more stable compound. When melted iron-carbon alloys of certain compositions cool to the solidification range, the carbon atoms are diffused as much as possible, that is, the carbon atoms in the melt are as far apart as they can get. When solidification begins, some of the carbon atoms must get together. There are two ways of combination, one without iron (graphite) and the other with iron (cementite). With iron atoms on all sides of the carbon atoms, cementite may form in accordance with the rule of availability. It is more difficult for the carbon atoms to get together to form the pure carbon or graphite than to combine with a portion of the iron to form cementite. Apparently in the absence of silicon or CO_2 or some other agent, there is no precipitation of the graphite from the cementite, but in the presence of certain other materials which act as catalizers, there is a tendency for cementite to decompose into iron and graphite. There can be no question that graphite is the more stable form under those conditions, so I think Dr. Honda himself could not argue that in certain irons graphite is less stable than cementite. This indicates that graphite forms whenever it gets an opportunity at a temperature below the solidification point. The thing I would like to urge here in general is that it is not necessary to accept all of Dr. Honda's assumptions in order to accept his main conclusions, and I for one wish to pay my respects to Dr. Honda for presenting this simplified form of the iron-carbon diagram before the Society.

METALS AND ALLOYS FOR INDUSTRIAL APPLICATIONS REQUIRING EXTREME STABILITY

By Jérome Strauss

Abstract

This paper records the uses of the most widely applied metals and alloys under conditions demanding permanence, either superficial, or otherwise, usually when exposed to the active destructive agents, in some form—corrosion, erosion and (chiefly) elevated temperature. These three agencies are first discussed in detail including the relationship of laboratory or semi-practical test data to the results obtained in actual practice, as well as the influence of manufacturing and fabricating methods and of the manner of application, upon the performance of a metallic material. The various alloys are described, grouped according to the element present in their composition in major proportion.

The copper group including the commercially pure metal, brasses, true bronzes, manganese bronzes, aluminum bronzes, nickel silvers, and others find wide application for exposure to atmosphere, sea water and natural waters. Pure aluminum and selected aluminum alloys have recently become of some importance in the chemical industries and for shipboard use. The pure metals tin, zinc, lead and their alloys are also discussed. Cast iron and simple carbon steel of structural grades still hold first position in respect to quantity consumed. But none of these materials are especially advantageous when exposed to very high temperatures, either alone, or combined with chemical attack. For such service much use has been made in late years of nickel alloys and of the highly alloyed special steels.

Introductory

IT MAY appear to those not so intimately acquainted with the details involved that the necessity for a paper of a general character upon such a subject, is questionable, especially in view of

A paper presented before the Western Metal Congress held in Los Angeles, January 14 to 18, 1929. The author, Jerome Strauss, a member of the society, is chief research engineer, Vanadium Corporation of America, Bridgeville, Pa. Manuscript received April 11, 1929.

the quite frequent summation in recent years of existing practice in these applications. Metallurgical-chemical-mechanical technology in the industries involved is moving forward, however, with such rapid strides that each year finds outstanding accomplishments not noted the year before. And hence a yearly review, though in part repetitive, could well serve the useful purposes of focusing attention upon the improvements of the year, tracing developments up to the time of writing and indicating the trend of further work. Perhaps a suggestion of other problems requiring investigative attack, would not be amiss.

The outstanding destructive agencies at work upon industrial metallic structures are chemical or electrochemical attack (corrosion), superficial mechanical attack (wear and erosion), high and low temperatures and temperature fluctuation (in their influence upon dimensions, stress and mechanical properties), and alternations or repetitions of stress. Frequently, two or more of these agencies are simultaneously involved. The fourth cause of deterioration noted here may justifiably be considered as beyond the scope of this title and will consequently be omitted from the discussion, except in reference to such applications as involve consideration of one or the other deteriorating influences when the latter is also of primary importance in the service to be rendered.

In the presentation of the subject matter, various methods may be employed, but the author prefers first to discuss these destructive agencies and the means at hand for evaluating their influence and of using the resultant information in a selection of materials for specific uses; following this, to indicate briefly the properties of the various materials in relation to a variety of applications. It is always to be borne in mind, however, that where the results of trials in actual identical practice are not available, surveys by tests of various descriptions may frequently serve only to indicate that two or three or four materials or combinations possess sufficient merit for more exhaustive study. Previous experiences and the results of laboratory studies are of value only in proportion to the approach of the conditions under which they were secured, to those involved in the intended use, additions, subtractions and modifications of factors or variables being of like importance. Further, the translation of small-scale operation data with these initial selections, into ultimate practice requires in addition, coordination with that

judgment that comes from varied experience and close observation. Fortunately or unfortunately, some commercial application and industrial development usually arrive in advance of data on the properties of new materials or "new" properties of old materials as well as the underlying principles of observed phenomena, and only conservative engineering judgment has prevented greater material losses.

Corrosion

The term "corrosion" is here used in its broadest sense, namely the breakdown of metals by conversion to their compounds, regardless of the medium responsible for the reaction or the auxiliary agencies that take part in producing the observed results.

There are a great many factors to be considered in dealing with the inter-behavior of metals or alloys with the media in which they operate, some concerned only with the metallic material and some only with the substances with which it is in contact. Stress in a metallic body, is of great importance and is a frequent accelerator of chemical attack even when uniformly distributed. When non-uniformly distributed, action of a medium is likely to be even more severe due to variation in electrochemical potential with variation in unit load. Similar results obtain, though in larger degree when metals are locally deformed cold and not subsequently annealed or tempered, as in severe straightening or bending as a final fabricating operation. On the other hand, uniform severe cold work, as in cold-drawn sheets, wires and cables, frequently results in no greater susceptibility to destruction than was the case with the same metal in a uniformly softer state. The accelerating effects just described are not always observed, some combinations of metals and corroding media yielding the same degree of deterioration regardless of stress value or distribution or of overstrain, but in the absence of irrefutable data giving proof of such condition, the removal of overstrain and the application of low stress will be assurance of enhanced service life. The description thus far has referred to static stress. If cyclic stress is considered, there is much recent experimental evidence on steels to show that even mildly corroding media such as tap water may reduce the allowable safe stress in the absence of corrosion by as much as 65 per cent, while in sea water the figure mounts to 80 per cent.

Pressure, quite aside from the stress induced by it, is of itself

often an accelerator of corrosive attack as has frequently been brought out in studies of oil-cracking-still materials. Increase in temperature is generally regarded as a powerful agent in increasing corrosion but as a word against dangerous generalization, it may be noted that some instances have been observed in which increase in temperature has resulted in a decrease in activity.

The contact of dissimilar metals or even of metals and some nonmetallic substances need be carefully considered. It is often true that serious differences in the solution pressure will exist between two steels of different compositions, or between two different aluminum-base alloys. In the latter, combinations are known in which one alloy was the electro-positive member of a pair when in one medium and the electronegative one in another. Concentration gradients and temperature gradients are likewise important and may result also in local electrochemical action; they are at times the result of design, or of precipitation of solids on metal surfaces from reacting liquid or gaseous media, or of methods of operation, or other causes.

So much has been written and spoken of the care to be exercised in the formulation of corrosion tests and the interpretation of results and their employment in designs and materials selection that it is questionable whether further comment is required here. It is perhaps sufficient to note again that even small deviations of laboratory procedure from operating conditions may result in a group of alloys displaying quite a different order of resistance to the corroding medium; the influential factors are sometimes the presence or absence of impurities, slight differences in temperature or concentration, nature of gases or vapors above a liquid medium or dissolved in it, impurities introduced into the medium (often by solution of the container), local defects and other inhomogeneities in large metallic masses, differences in surface condition between specimen and article as used, etc. Generally it is the acceleration or intensification of one or more variables (sometimes intentional, sometimes unintentional) that produces a disparity. Whether with or without acceleration, the exposure period in testing is a prime factor too often neglected. When films form and are protective the corrosion rate decreases rapidly; when they form and are porous to the medium the corrosion rate may either be steady or only slightly reduced, or it may rise due to electrochemical stimulation; when

such films expand or contract strongly or for other reasons break away from the parent metal, more complex conditions exist. In every case then, only a full insight into both tests and service can produce sound decision. In chemical processes the degree of contamination of the product through solution of the equipment may be a deciding factor. On the other hand, suitable contamination of a medium, where permissible, may serve as a corrosion inhibitor.

Wear and Erosion

Of the destructive agencies considered here, no doubt the least is known concerning the mechanical attrition of metallic surfaces. Here again test results fail to provide a reliable indication of service life unless the test itself (or group of tests) duplicates the conditions of use. Where two surfaces are in contact, the properties of the metals at the surface (hardness, toughness, microstructures, mechanical finish) as well as the properties of the metals just below the surface and also the speed, magnitude and character of loading, etc., along the contact area, are of importance. The nature of the metal below the surface is of especial interest in coated or shallow-hardened objects. In general, metal to metal contact not being desired in moving parts of machinery, etc., the surfaces are kept apart by suitable films and the above considerations enter only when film breakdown occurs; they are, of course, important in various cutting and forming operations. When the action involved is impingement of relatively small masses of nonmetallic or metallic solid material, or of liquids carrying such masses, upon a relatively larger metallic mass and the latter is the part requiring replacement when deteriorated, not only its properties as noted above are to be considered but also the surface condition, shape and physical properties of the impinging material, the nature of the liquid, and the speed and manner of impingement. Both hardness and toughness in the metal are important, to a degree and of a character not yet well analyzed but probably in some instances either one being of greater importance than the other.

Effect of Temperature

A far greater amount of data exists on the mechanical properties of metals at elevated temperatures and yet the total but a small fraction of what the designing engineer truly requires for

the construction of high temperature steam and chemical plants, furnace equipment and the like. Withal, he is continually moving on to higher pressures and higher temperatures with designing built upon meager data, but some experience and often much conservatism.

Information acquired during the last few years and being developed at present by painstaking research is largely upon the properties determined by tension testing of small sections. Two methods of testing are in vogue, the one resulting from the shortcomings of the other. One may determine proportional limit, yield point, tensile strength, elongation and reduction of area as in room temperature testing but in so doing it will be observed that as the testing temperature increases the results become more and more dependent upon the speed of testing, the strength properties decreasing progressively as the speed of load application diminishes. This has led to the measurement of deformation at fixed load and temperature over long time periods; the data secured in this manner is of unquestioned value if properly applied to each design need but the time and cost of assembling it as well as the fact that few tests of a duration comparable to that of anticipated service life have been recorded, make the correlation with some shorter method of examination a much sought goal.

It has been noted that up to about 900 to 1100 degrees Fahr. the proportional limit as determined in tests of high accuracy requiring a few hours for their completion, closely approximates the maximum load that can be sustained with very little deformation throughout long life. Such correlations, based as they are upon small numbers of tests and being admittedly only approximations, must be employed with due caution. Other brief methods of reaching design data have been proposed based upon measurement of rates of flow, after flow has become steady in the early stages of long-time tests; being formulated, however, upon assumed continuance of the steady rate and as permissible rates must vary with types of applications, these methods are likewise susceptible of much misuse. Another suggested method consists in the determination of the time required to produce rupture at various stresses and then, from the resultant curve of stress plotted against time, obtaining the apparently asymptotic value of strength (with time of the order of several thousand hours); a design value is then deduced

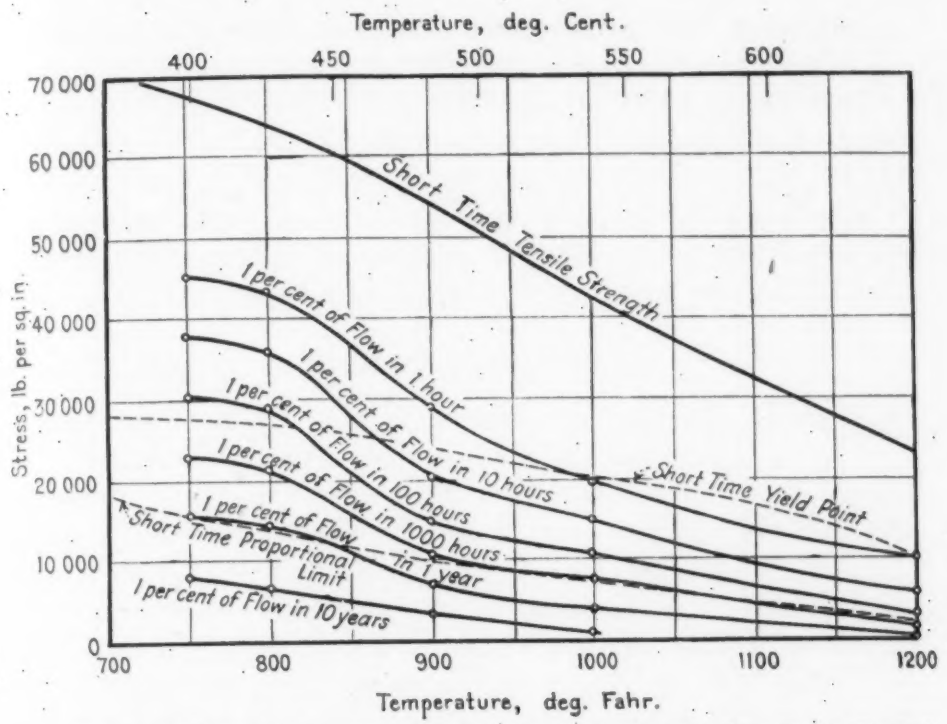


Fig. 1—"Iso-Flow-Rate" Curves for Carbon Cast Steel with Short-Time Tensile Strength, Yield Point, and Proportional Limit Curves Drawn in for Comparison.

by the application of a factor of safety. This may be applicable in certain types of structures but it can quite conceivably result in over-designing or under-designing unless checked by some other means of measurement. A very excellent method of employing flow test data may be illustrated by a family of curves taken from a recent publication of Spring and Kanter; using such a plot (Fig. 1) it is possible to select design stresses for any given permissible deformation. This is essential since the allowable movement under loading in such structures as turbine blading and blade fastenings and certain parts of valves must of necessity be far less than that of supporting members in furnaces and the like. It is important to note here that there are temperature ranges in which some metals are brittle and others in which the same metals are ductile and that this will influence the manner of failure should unforseen conditions bring about sufficiently high local stressing or suitable temperature. Also, in a tubular member, the allowance of too great a flow rate may automatically result in premature failure since increase in size and simultaneous decrease in wall thickness both introduce increased unit stress.

Extremely low temperatures also create material problems since many steels, for instance, rapidly develop brittleness with decrease in temperature and with an increasing number of industrial processes involving such operations, data of this sort are needed. Fortunately its accumulation is relatively a simple matter, not requiring either the delicate and elaborate equipment or involving such controversial questions of interpretation as are encountered when dealing with temperatures well above atmospheric.

General Conditions

Ultimate selection of materials involves, obviously, considerations other than those just enumerated. Ability to secure a material in the most suitable form for a specific purpose, the cost of raw material and the cost of such fabricating operations as are required to put it into the desired final form, are of prime importance. This in turn involves the ability to perform, and the influence of, such operations as casting, forging, rolling, welding, machining, etc. And at times other mechanical properties need be considered; such, for instance, as ability to serve as a bearing surface for a moving part. Final selection must then represent the best balance of chemical and physical properties, fabricating ability and the economics of the structure and the process or service.

Copper and its Alloys

Historically and commercially this group form the oldest of corrosion-resisting metallic materials. While commercially pure copper is exceptionably permanent in contact with a great variety of corrosive liquids and gases or vapors, it owes much of its extensive use to its great ease of fabrication, both hot and cold. It is available in a high degree of purity in every common wrought form. In the cast state it may be had from many sources in almost as high a degree of purity if correspondingly high electrical conductivity is not demanded; should the latter be essential, then the results are attainable chiefly by the use of alloys or compounds of boron as scavenging and degasifying agents; the purity is as high as in the wrought product but the sources of supply become limited. At appreciably elevated temperatures copper is of little value due to rapid reaction if exposed to the atmosphere or to furnace gases, as well as to low load-carrying ability. At low temperatures, how-

Table I

Tension Tests of Some Metals at Room and Liquid Air Temperatures

Material	Table No.	Alloy No.	Room Temperature Properties: Yield Point Lbs. Per Sq. In.	Tensile Strength Lbs. Per Sq. In.	Elong. % 2"	Red. Area %	Boiling Liquid Air Properties: Yield Point Lbs. Per Sq. In.	Ten. Str. Lbs. Per Sq. In.	Elong. % 2"	Red. Area %
Cold Rolled Copper.	II	2	43,300	45,800	16.0	55.9	50,400	53,000	19.0	55.4
Annealed Copper ...	II	2	12,700	32,600	58.0	44.3	18,300	50,400	63.0	68.2
Rolled Tin Bronze..	II	4	57,000	61,600	36.0	65.4	71,800	93,200	56.3	58.0
Cast Tin Bronze...	II	7	18,600	40,200	31.3	36.7	30,500	45,300	15.3	24.6
Cast Al. Bronze....	II	17	29,000	70,300	33.3	32.6	42,600	85,900	24.9	22.4
Forged Al. Bronze..	II	16	47,900	88,900	45.2	46.9	84,900	112,500	38.4	42.1
Rolled Si Bronze...	II	22	34,100	67,500	49.0	45.2	39,700	88,100	51.3	41.3
Rolled Leaded Brass	II	36	24,400	35,600	17.0	22.2	32,100	42,800	13.0	19.0
Rolled Naval Brass.	II	34	28,800	57,100	47.4	50.5	37,200	81,100	48.3	48.4
Cast Mn. Bronze ...	II	24	31,600	68,500	39.2	32.5	40,800	82,700	26.3	25.8
Cold Rolled Ni Silver	II	37	69,300	73,800	21.5	54.3	80,500	93,700	35.5	62.6
Annealed Ni Silver..	II	37	29,500	64,900	46.8	62.3	38,200	83,000	56.8	69.5
Cast Ni Silver	II	37	16,800	40,300	40.5	39.2	23,900	49,200	32.7	30.2
Cast Al. Alloy	III	11	7,600	18,100	8.8	10.2	8,100	17,800	7.0	7.3
Cast Al. Si. Alloy..	III	7	9,200	17,300	4.9	5.2	9,600	19,600	3.7	4.7
Duralumin	III	4	35,400	57,800	26.5	27.0	42,700	71,800	28.0	28.7
Hot Rolled Nickel..	IV	1	54,400	93,400	32.7	45.5	81,800	122,500	46.7	57.2
Hot Rolled Monel..	IV	4	44,100	91,500	45.7	66.9	70,800	135,500	53.5	67.0
Annealed 0.4 C. Steel	...	..	45,800	79,400	30.8	49.0	114,100	139,400	7.3	7.1
Treated 0.4 C. Steel	...	..	76,900	104,400	25.0	61.3	150,200	160,400	9.8	9.4
Annealed 3 Ni. Steel	V	7	57,000	79,400	31.5	57.2	111,000	116,600	6.8	6.7
Treated 3 Ni. Steel	V	7	76,400	98,800	27.3	65.9	132,900	148,700	15.5	17.6
Annealed 16 Cr. Iron	VII	19	42,800	75,100	34.0	70.8	123,200	145,700	14.0	14.3
Treated 14 Cr. Steel	VII	15	212,200	225,800	9.8	37.3	210,800	259,700	2.8	*3.5
Treated 20 Cr. Steel	VII	25	51,900	91,900	24.0	59.3	99,200	99,200	.0	.0
Annealed 26 Ni. Steel	VII	29	52,900	99,800	38.3	52.1	157,800	205,000	11.5	9.7
Annealed 21 Ni., 8 Cr. Steel	VII	39	48,300	107,600	44.3	60.1	102,800	180,200	25.3	21.5

*Broke in 3 places.

ever, it retains its ductility remarkably well and has found numerous applications; chiefly, as tubing. The properties at liquid air temperature appear in Table I.

Table II lists typical, commonly employed alloys in which copper is the major component. In addition to the nominal composition (other elements are frequently present in minor amounts due to raw material and processing limitations) some measure of their physical character is given by the notation of room-temperature tensile strength. While other properties such as elastic strength, hardness and ductility are of interest and importance, space limitations obviously prevent their inclusion. For this and other information reference must be had to more detailed presentations. The strength ranges or variations quoted in Table I are in general largely the result, in the cast alloys, of the rate of solidification and subsequent cooling and, in the wrought alloys, of the degree of cold working and of the temperature and time in annealing cycles; in a few instances notably the aluminum bronzes, they may also result from quenching and tempering treatments.

Table II
Copper and Copper Base Alloys

Designation	Condition	No.	Nominal Chemical Composition Cu	Sn	Zn	Fe	Al	Pb	Mn	Ni	Si	P	Tensile Strength 1000 Lbs. Per Sq. In.
Copper	Cast	1	99.8	..	..	..	..	..	..	..	..	..	25-30
	Wrought	2	99.9	..	..	..	..	..	..	..	..	..	30-50
Tin Bronzes													
	Wrought	3	90	10	..	..	..	..	..	..	..	..	50-100
	Wrought	4	95.7	4	..	..	..	..	..	..	..	0.3	50-150
	Cast	5	80	20	..	..	..	..	..	..	..	Tr.	30-35
	Cast	6	90	10	..	..	..	..	..	..	..	Tr.	30-35
	Cast	7	88	8	4	..	..	..	..	..	..	..	35-45
	Cast	8	88.5	11	..	..	..	0.3	..	..	..	0.2	30-40
	Cast	9	86	9.5	2	..	..	2.5	..	..	..	..	30-40
	Cast	10	86	12	1.5	..	..	0.5	..	..	..	..	35-45
	Cast	11	80	10	..	..	..	10	..	..	..	..	25-35
	Cast	12	78	8	..	..	..	14	..	..	..	..	25-30
Aluminum Bronzes													
	Wrought	13	86	..	..	4	10	..	..	..	..	..	85-110
	Cast	14	89	..	..	1	10	..	..	..	..	..	70-80
	Cast	15	90	..	..	..	10	..	..	..	..	..	70-75
	Wrought	16	88	..	..	3	9	..	..	..	..	..	70-130
	Cast	17	88	..	..	3	9	..	..	..	..	..	70-80
	Wrought	18	92	..	..	..	8	..	..	..	..	..	70-130
	Wrought	19	82	..	..	8	10	..	..	..	..	..	65-80
	Cast	20	82	..	..	8	10	..	..	..	..	..	70-90
Silicon Bronzes													
	Wrought	21	98.5	1.5	..	..	..	..	..	..	0.05	..	35-70
	Wrought	22	94.5	..	..	..	..	..	1	..	4.5	..	55-140
	Cast	23	94.5	..	..	..	..	..	1	..	4.5	..	50-60
Manganese Bronzes													
	Cast	24	57.7	0.6	39	1.2	1.0	..	0.5	..	..	..	65-80
	Cast	25	68	..	22	2	4	..	4	..	..	..	85-95
	Wrought	26	68	..	22	2	4	..	4	..	..	..	100-120
	Wrought	27	58.7	..	40	1.0	0.2	..	0.5	..	..	..	70-90
	Wrought	28	65	..	23	2	..	..	..	10	..	..	50-160
	Cast	29	60	..	38	..	0.4	..	0.8	..	0.8	..	75-90
Brasses	Wrought	30	80	..	20	..	..	..	..	..	..	..	40-80
	Wrought	31	65	..	35	..	..	..	..	..	..	..	45-80
	Wrought	32	71	1	28	..	..	..	..	..	..	..	45-70
	Cast	33	62.5	1	36.3	..	0.2	..	..	..	..	..	30-35
	Wrought	34	61	1	38	..	..	..	..	..	..	..	50-90
	Wrought	35	60	..	40	..	..	..	..	..	..	..	45-80
	Wrought	36	61	..	36	..	..	3	..	..	..	..	35-60
Nickel Silvers													
	Wrought	37	64	..	18.7	..	..	..	0.3	17	..	..	60-150
	Wrought	38	64	..	17.7	..	..	1	0.3	17	..	..	55-120
	Cast	39	57	2	16	..	..	6	..	19.5	..	..	30-40
	Cast	40	54	..	20	..	..	..	0.3	25	0.7	..	70-80
Cupro Nickels													
	Wrought	41	70	1	..	..	..	..	..	29	..	..	60-120
	Wrought	42	68.5	..	..	..	..	..	4	26	0.5*	..	70-150
	Wrought	43	59	..	..	..	..	..	1.5	35	..	4†	80-110
	Wrought	44	75	..	5	..	..	..	..	20	..	..	50-85
	Cast	45	63	..	..	..	..	..	..	29.5	0.5*	..	50-60
	Cast	46	67	..	..	..	..	..	1.5	29	0.5*	..	60-70
	Cast	47	55	..	5	..	..	..	..	35	..	..	55-80
Manganese Copper													
	Wrought	48	96	..	..	..	..	..	4	..	..	..	40-70

Note {* Si + C; † Cr}

Aside from the very general use of copper, silicon bronze, brasses and nickel silvers for resistance to the elements in atmospheric exposure, most of the copper-base alloys have had extensive application as containers and mechanisms for handling sulphuric, sulphurous and hydrochloric acid (cold, dilute) and their salts as well as many other acid and neutral salts, crude acetic acid, mine waters, alkalis, tanning liquors and dyewood extracts; they have likewise been extremely serviceable in contact with sea water and sea-water spray in general marine service. In many cases, however, the presence or absence of air in contact with liquid and metal may appreciably alter the results obtained (this is especially true with hydrochloric acid).

Those tin bronzes which are provided in wrought form are cold-worked exclusively, hot working being unsuccessful; due to this and other manufacturing limitations central integrity is not always to be had in sizes above about 2½ inches. The tin bronzes with and without lead, because of the properties of their micro-constituents, form a very important group of bearing alloys; some of the harder types such as Nos. 8, 9 and 10 meet a demand requiring both strength, wear-resistance and bearing qualities as in worms and worm wheels. The high tin alloys (Nos. 3 and 6) are in especial demand for resistance to corrosion as are likewise the cast alloys containing much lead. In the latter group, part at least, of the serviceability is due to physical effects of the lead in decreasing porosity since obviously porous castings, as a result of greater exposed surface and less actual metal thickness must be less durable in corroding media than are those which are thoroughly sound. Bronzes Nos. 7 to 12 inclusive are among the most desirable of the founder's alloys and readily yield sound castings in skilled hands. The aluminum-bronzes are noted for their combination of very high strength and high ductility, their response to heat treatment and the small difference in properties between the cast and forged states. Those with higher aluminum (duplex alloys) possess good bearing qualities and have had much success as worms or worm wheels. The high shrinkage of all the commercial alloys and their tendency to form aluminum oxide (dross) in melting and pouring make the production of sound castings not a simple matter. It is, however, being successfully accomplished by many founders while extremely sound ingots for forging have been pre-

pared as large as 15,000 pounds. The corrosion resistance of this group is of a very high order.

The silicon bronze represented by Nos. 22 and 23 is one of the newer alloys and has been widely applied. It possesses the desirable advantages of corrosion resistance comparable with that of the tin-bronzes, generally better mechanical properties and a price advantage in comparison with many copper-base alloys. But it is not an especially easy alloy to handle in some manufacturing operations, such as casting in large or irregular sections or in moderately large ingots; with increased study and use, however, these difficulties are decreasing.

The manganese bronzes and brasses are not generally as resistant to corrosion as the copper alloys thus far treated. They do, however, have a price advantage. The brasses, especially Nos. 30 and 31, are readily formed into tubing. Alloy No. 32 is made into tubing, although with some difficulty, but it has shown so much better resistance to sea water and fresh water corrosion than most of the others that it is quite extensively used as condenser tubes; alloy No. 30, however, is entering this field to an increasing degree. Brass tubing is also used in oil refinery condensers and heat exchangers. Manganese bronzes of the type represented by No. 24, due to the combination of low cost, moderately good sea water corrosion resistance and strength, will be found in practically every ship propeller. These alloys are similar to aluminum bronze in respect to casting difficulties (though to a lesser degree) but the propeller form is so simple that the difficulties are minimized. Alloy No. 29 is a newcomer in the field but is finding wide application, in both the cast and forged condition, as wearing parts such as gears and worms. It is well to note that manganese bronzes containing much aluminum are subject to grain growth and the development of brittleness if hot-worked much above 1300 degrees Fahr.; also that all of these two-phase bronzes have a range of heat fragility at a somewhat lower temperature (about 600 to 900 degrees Fahr.) which is wider when much aluminum, the chief strengthening component, is present. It is also of interest that these alloys stand alone in their serviceability in constructions for handling formaldehyde.

The nickel silvers find their largest uses as parts for atmospheric exposure and as resistance elements operating at low temperatures. They do, however, possess exceptional surface stability

in contact with sea water and sea water spray. Comparatively high cost bars them from many fields as does also a lack of real ease of manufacture. Sound castings are not the rule but those alloys containing lead or silicon and manganese have advantages in this respect while the latter alloy (No. 40) possesses very high strength as well.

None of the copper alloys thus far mentioned are of special merit for load-sustaining or oxidation-resistance at somewhat elevated temperatures. In some cases immediately above room temperature, but generally between 450 and 550 degrees Fahr. the strength properties begin to decrease and as the proportional limits are initially low, it is probable that at 800 to 1000 degrees Fahr. their load-sustaining capacity is close to zero. Many of the cupronickels on the other hand have properties more similar to those of the higher nickel alloys and show quite moderate changes in load-sustaining ability as the temperature is raised to 1000 degrees Fahr. In addition, No. 43 is claimed to possess good grinding properties, a quality not generally characterizing alloys of copper and nickel. Most of these alloys have found application in valve parts for steam service and the like. All of them are forgeable if carefully handled, but they are notably hot-short. The manganese copper is quite permanent when exposed to corrosion, to oxidation and to moderately elevated temperatures and has been much used for firebox plates and stays, and for turbine blading, although being rapidly displaced from the latter field by the iron-chromium alloys.

It should be noted that all of the copper-base alloys may be brazed and soldered. For aluminum bronzes special fluxes are needed for brazing and copper plating is of much aid prior to soldering. All of the alloys, however, the manganese bronzes and duplex brasses in particular, are subject to the development of intercrystalline brittleness in this operation depending upon the temperature and the stress in the metal. These same duplex alloys also, it may be noted, frequently yield to a form of very damaging selective corrosion, known as dezincification, the mechanism of which is as yet not fully agreed upon.

Aluminum and its Alloys

Aluminum has found its way into the chemical industries more than have its alloys. This appears to be not so much a matter of

corrosion resistance as the result of a demand for wrought material of no great strength rather than for strong materials or for castings. Thus riveted and welded vessels built of plates and shapes have become common for handling pure acetic acid of various concentrations and for the preparation of a great variety of foodstuffs. Lately, tank cars for the transportation of acetic acid have been successfully operated with tanks and other parts of aluminum. Nitric acid in certain ranges of concentration and temperature does not appreciably attack aluminum and there has been a rapidly growing effort to use the metal in the preparation of this acid. These applications have largely employed commercially pure (99 per cent plus) aluminum and in this grade the resistance to corroding media increases rapidly with decrease in the content of the impurities copper and iron.

Pure aluminum also corrodes at a very slow rate when in contact with sea water spray but for applications of this nature stronger materials are demanded. The wrought and cast high strength alloys as well as some of intermediate strength, have come into use for such purposes (alloys Nos. 3 to 11) and while all are protected when possible, with paint or varnish coatings, the alloys high in copper require the protection more than do the others. Those of the alloys that are best suited to unprotected exposure are Nos. 3, 7, 9, 10 and 11. Alloy No. 8 requires special treatment in the molten state for the development of its high strength and ductility and is likely to produce somewhat erratic physical properties while alloy No. 11 yields shrinkage defects in other than experienced hands. Intercrystalline corrosion is commonly found in the high copper heat treated alloys under some conditions of manufacture, heat treatment and exposure but may be minimized by suitable production methods. Alloys such as No. 3 appear not so subject to this type of failure but absolute immunity is obtained as well as greatly enhanced resistance to corrosion from a product of the type of No. 4 (duralumin) carrying a pure aluminum coating of the No. 1 composition.

Aluminum-alloy coated (calorized) iron and steel articles have also been subjected to severe service in the form of pyrometer tubes, oil still tubes, furnace parts and the like and while providing better service than similar parts of ordinary iron and steel which they replaced, are largely giving way to the highly alloyed steels.

Mechanical properties of aluminum alloys at elevated temperature are such that only limited use can be made of a limited number. The principal application has been in internal combustion engine pistons, where the actuating motive was low unit weight. In the softer aluminum alloys strength properties decrease rapidly with increase in temperature, while in the harder ones they hold up quite well only to between 400 and 500 degrees Fahr. At 800 degrees Fahr. load-carrying ability is low, generally less than 2000 pounds per square inch, and in the softer alloys practically zero.

Aluminum and its alloys present no great casting difficulties and those that are workable may be easily wrought once the relation of temperature and permissible reduction are understood. The alloys may be soldered but the solders required are of such nature that they produce deterioration of the aluminum by electrolytic corrosion and should not be employed when corroding media are in contact with the metal. Welding by skilled personnel provides a suitable method of joining; on heat treated parts, heat treatment should be performed after welding, else there is electrolytic corrosion in which the parent metal, the weld metal and the metal between them that has been softened by the heating, all participate.

Minor Metals and Alloys

The metals tin, zinc, lead, cadmium, silver and chromium and some of their alloys have found minor uses. The major use for tin is in the form of a pure or alloyed coating (with 75 per cent lead, as in terne plate) on iron and soft steel. Both coatings are employed for structures or articles subject to atmospheric exposure while the pure tin-surfaced material finds a very wide outlet in the "tin" cans for food products. The latter application results from the immunity of tin to the attack of most organic acids in the absence of air or other oxidizing substances. These uses all involve thinly-coated metals. But heavier coatings are much used over copper and mild steel in vessels, piping, etc. employed in the food manufacturing and drug industries as tin is only slightly attacked by food acids when air is present and the period of contact is short, and is further quite resistant to most pharmaceuticals. In many electrolytes, however, tin is electronegative to iron and steel and hence with thinly coated articles especially, slight discontinuities in the coating may prove very disastrous. Due to high cost, tin in

massive form is very little used but is still to be found as block tin tubing in many forms of apparatus for the preparation of distilled water, the metal being regarded generally as unequalled for this purpose by any other of equal or less value. Tin may contain impurities that make it susceptible to intercrystalline corrosion under the action of some corroding agents, while severe strain also frequently results in rapid decay.

Zinc finds its way into the types of service considered in this paper, largely as a coating on irons and soft steels, for improving their life when exposed to the atmosphere. This it accomplishes to a marked degree, though not so strikingly in those industrial regions where the air is contaminated with large amounts of sulphur compounds. It has also found some use in building construction in the from of hard rolled pure sheet (weather strips, roofing shingles, gutters, etc.) and likewise forms the shell and negative electrode of the common dry cell. This metal also is subject under some conditions of composition, stress and corrosion, to intercrystalline attack.

Lead is still the most used metal in sulphuric acid production, especially in the so-called contact process although its solubility under these conditions varies appreciably according to its purity and the temperature, concentration and purity of the acid in contact with it. It likewise finds application in many other processes where free sulphuric acid is present in other than low concentration, such as petroleum refining and chromic acid manufacture. A further group of uses include the handling of many of both the organic and inorganic salts of ammonia while another consists of its employment either pure or alloyed with other metals as anodes in various electrolytic processes. In spite of all these uses, lead itself is not inherently adapted to structural employment because of its poor strength and rigidity. This results in alloying it with from 2 to 18 per cent of antimony which adversely affects its resistance to the attack of many media, or in applying it as a coating to steel as in coated sheets or sheet construction or to cast iron as in piping, valves, pumps, etc.

Cadmium has recently come into extensive use as a coating for iron and steel exposed to the atmosphere and to both fresh and salt waters. In the latter it not only affords electrolytic protection at coating discontinuities as does zinc but is directly very much less soluble than zinc and hence has far greater life. Most of the

Table III

Aluminum and Aluminum Base Alloys

No.	Form	Nominal Chemical Composition							Tensile Strength 1000 Lbs. Per Sq. In.
		Al	Fe	Si	Mn	Cu	Mg	Ni	
1	Wrought	99.8	..	..	..	..	..	..	9-10
2	Wrought	99.2	0.5	0.2	..	0.1	..	..	12-22
3	Wrought	97.8	0.6	1.0	..	..	0.6	..	25-65
4	Wrought	93.6	0.6	0.3	0.6	4.5	0.4	..	50-85
5	Wrought	94.1	0.6	0.3	..	5	..	..	40-75
6	Cast	94.2	0.6	0.8	..	4.4	..	..	25-45
7	Cast	94.2	0.6	5	..	0.2	..	..	17-20
8	Cast	85.7	0.6	13	0.5	0.2	..	..	25-30
9	Wrought	97.5	0.5	0.3	1.5	0.2	..	..	14-26
10	Cast	97.0	0.5	0.3	2.0	0.2	..	..	18-20
11	Cast	97.4	0.5	0.3	0.8	1.0	..	..	18-20
12	Cast	86.2	0.5	0.3	1.0	12	..	..	20-25
13	Cast	88.3	1.2	0.3	..	10	0.2	..	25-30
14	Cast	91.7	0.5	0.3	..	4	1.5	2.0	25-45

coating work has been by electrodeposition. Spraying, however, has been very successful but the deposit for a given purpose must be heavier while the toxic effects of the vapors, more serious than in the case of zinc, must be guarded against. Electrodeposited cadmium coatings have been shown to very materially raise the resistance of steels to alternating stresses when exposed to fresh water.

Chromium in the past few years has made rapid strides as a coating on many metals. It has been employed for its inherent resistance to corrosion in the atmosphere at low temperatures, to oxidation at quite elevated temperatures and to wear and abrasion. Depending upon the conditions of its employment, it may have to be underlayed by a coating of another less corrosion-resistant metal of not much different solution potential, such as nickel, so that the effects of coating imperfections may be minimized, or it may have to be backed by rigid material when subject to heavy loading so that it will not be greatly deflected and crack. These cautions are, of course, intimately related to the conditions of deposition and will no doubt be subject to modifications as improved processes and technique are developed.

Other metals also find some use in minor quantities such as, for instance, the alloys of magnesium for aircraft engine pistons and coatings of silver for springs exposed to some corroding media and for apparatus for use with certain concentrations of hydrochloric acid and other reactive substances.

Table IV
Nickel and Nickel Base Alloys

No.	Condition	Nominal Chemical Composition Ni	Cu	C	Mn	Fe	Cr	Si	Tensile Strength 1000 Lbs. Per Sq. In.
1	Wrought	99.2	0.2	0.1	0.1	0.4	..	..	60-130
2	Wrought	97.2	0.2	0.1	2.0	0.5	..	..	65-130
3	Wrought	94.6	0 2	0.1	4.5	0.6	..	..	65-130
4	Wrought	67	28	0.5*	2.5	.2	..	..	65-140
5	Cast	67	27.5	1.5*	2	2	..	..	65-75
6	Cast	60	26	0.5*	1	1.5	..	Sn 11	....
7	Wrought	70	26	0.5*	3.5	..	..	..	55-140
8	Wrought	85	..	..	..	..	15	..	80-110
9	Wrought	80	..	..	..	..	20	..	90-120
10	Cast	68	..	.1	1	7	21	2	55-65
11	Cast	65	..	1	2	16	15.5	0.5	....
12	Cast	60	..	0.5	2	23.5	13	1	....
13	Cast	61	Al 2	1	1.5	10	21	3.5	....
14	Cast	66	Al 1	1	1	12.5	16	2.5	....
15	Cast	58	W 4	0.5*	1.5	24	12	..	....
16	Wrought	58	Al 3	0.1	2	15	21	0.9	....
17	Wrought	53	Al 3	0.3	2	27	14	0.7	....
18	Cast	63	8	0.5	Mo 4	W 2	21	1.5	80-90

*C + Si

Nickel and Nickel Alloys

The alloys rich in nickel commonly used for resistance to corrosion and to high temperature appear in Table IV. The use of these alloys has been limited to a great degree by their first cost in spite of excellent proof that in some services overall costs would be lower than with initially less-expensive alloys.

The nickel, nickel-manganese and nickel-copper types have been somewhat more serviceable than brasses and bronzes in exposure to the atmosphere, fresh water, sea water, sulphuric acid and cold hydrochloric acid. While with sulphuric acid containing oxidizing salts they do not show as great resistance as the bronzes, on the other hand pure nickel in contact with acetic acid and an oxidizing agent such as air is far superior to copper under the same conditions. Pure nickel has been much used for resistance to the organic acids in foodstuffs in the form of equipment for kitchen use and for the food products industry.

Nickel and high-nickel alloys are not serviceable in hot gases containing appreciable amounts of sulphur compounds and heating them in such atmospheres prior to forging will result in hot-shortness; these alloys are also not especially suited to hot atmospheres containing much carbon monoxide. With the exercise of care and the use of special methods, these alloys may be welded, while braz-

ing is rather readily accomplished. The pure nickel is one of the few metals and alloys that have been found to silver-solder without the development of intercrystalline brittleness even when stressed or heated slightly beyond the required temperature.

Alloy No. 6 differs somewhat from the other nickel-copper alloys and has been widely employed for valve parts in high pressure steam service where it, like alloy No. 43 of Table II, possesses the advantage over nickel and monel metal of better grinding properties. The latter two, nevertheless, enjoy some use in high-pressure steam service not only in valves but in blades and other parts of turbines. The nickel-manganese alloy No. 3 seems to have given the best service as anchor bolts for the refractories in marine boiler furnaces operating at high temperature. Practically all of the alloys of Table IV serve admirably in contact with alkaline substances such as hot and cold solutions of ammonia, carbonates and caustic alkalis; also with the latter when in the fused state.

The iron-free nickel-chromium alloys are chiefly high temperature electrical heaters and as such show excellent resistance to oxidation up to 1800 to 2000 degrees Fahr., although in mechanical properties at these temperatures they are excelled by many iron-base alloys. These two combinations are somewhat difficult to work and command a rather high price.

The nickel-chromium-iron alloys have seen a small amount of service in resistance to acid and chemical attack at room temperature or only slightly higher. The nickel-chromium-copper alloy with tungsten and molybdenum (No. 18) is used for this purpose only, being especially suited for sulphuric acid and sulphuric-nitric acid mixtures. Lately it has been applied to the manufacture of parts of acid pumps; price, however, rules it out of many possible fields.

A great proportion of the cast parts of both batch and continuous furnaces for heat treatment of metals (including carburizing boxes) enameling and other ceramic-plant operations, ore preparation processes, etc., operating at about 1475 to 2000 degrees Fahr. and sometimes higher are made of alloys of the nickel-chromium-iron types such as Nos. 10 to 14 of Table IV, in spite of their high cost and of strong competition from higher-iron alloys. Their use in this service has resulted from excellent resistance to the attack of air and furnace gases at high temperature combined with ex-

cellent load carrying ability. This development has grown in the face of great casting difficulties not yet fully overcome. In fact such failures as have been experienced are ascribable in far greater degree to manufacturing shortcomings than to improper selection of composition. The same facts probably account for the absence of truly reliable information as to their mechanical properties at room or high temperatures.

An alloy of the type of No. 15 has been much used in European adaptations of ammonia synthesis processes while the wrought alloys as wire and ribbon have been used as resistances and heaters operating at temperatures up to about 1200 to 1400 degrees Fahr.

Iron and Iron Alloys

Cast Iron

One of the two most common materials of construction of the chemical industry is cast iron—the other being mild steel. Both find such wide employment not because of unusual corrosion resistance but rather because of low initial cost, combined with, in the case of cast iron, great ease of casting and the ability to produce readily forms of considerable complexity, and in the case of mild steel ready workability into all fabricated forms. Extreme stability is often sacrificed for these advantages the problem being for each application, careful weighing of the cost of less resistant materials with their availability in desired forms plus the cost of replacements, failures and service interruptions, against the cost of long-life materials in suitable form with their process advantages but added investment charges.

For exposure to inorganic acids, cast irons should preferably be of high total carbon and low combined carbon content. This obviously means low silicon content in plain irons and this is desirable, especially for oleum and fuming acid as free SO_3 rapidly oxidizes the silicon selectively. Resistance to corrosion in these acids increases rapidly as temperatures are lowered and concentrations are raised. Cast iron serves also for mixed acid and some of the organic acids. For alkalis, irons with low total carbon content (and the greater portion of that graphitic) are desired in order to secure long life; but low silicon content also favors this objective of alkali-resistance and as the two requirements are opposed metallurgically,

1 to 3 per cent of nickel is frequently added not only to insure grayness in the presence of low silicon but for its direct result upon resistance to attack. Vessels of such composition are used for hot and cold, dilute and concentrated as well as fused caustic alkalis. High heating of the fused substances, however, (to 570 degrees Fahr. frequently) causes rapid deterioration of the iron unless careful engineering has avoided any chance of local overheating.

Cast iron finds other fields of usefulness at slightly elevated temperatures. In pistons and cylinders for internal combustion engines, it is still of prime importance; low carbon content (about 3.0 per cent) and low phosphorus (about 0.20 per cent) are generally preferred for this service. Low phosphorus irons (some even as low as 0.03 per cent) are also desired for steel plant ingot molds, while the manganese content is in some cases raised to the vicinity of 1.0 per cent. In steam plant construction much use has been made of cast iron up to about 435 degrees Fahr., the permissible unit stresses dropping rapidly as the temperature increases to this maximum; these applications have been largely valve bodies, fittings and turbine casings.

Malleable iron, an initially white cast iron of low carbon and low silicon content (about 2.5 and 0.7 per cent respectively) rendered ductile by suitable thermal treatment, while not much used under corrosive attack, is a common material for steam power plant auxiliary appliances and is serviceable at somewhat higher loads and slightly higher temperatures than gray cast iron.

For all of these applications it must be noted, not only chemical composition of the iron is important but likewise the metallurgical practice involved in the production of castings which greatly influences the degree of porosity and the fineness of the graphite particles; the rate of cooling during and after solidification is also important not only for its influence upon structure and physical properties but because of the residual casting strains which may be released during machining with change in shape of the parts, or during use especially if temperature rise or change is involved, unless suitable heat treatment has been applied subsequent to removal from the mold.

Closely related is the increase in bulk (increase in dimensions and decrease in density) of some cast irons (known as "growth") accompanying frequent temperature changes of large magnitude.

The amount of the volume change depends upon composition, metallurgical features of production, and the time-temperature cycles and their frequency. It has been reported that white irons change less than those initially gray and that the most suitable is one of about 3 per cent carbon, very low silicon, and not much more than traces of other elements. This iron, like washed metal, is unmachinable, yet quite tough, and is suited to the manufacture of many furnace parts.

Cast irons also find applications for wear resistance, either as especially dense castings of fine graphite distribution resulting from the addition of special agents inhibiting the growth of carbon particles such as calcium silicide, or by addition, with suitable adjustment of silicon and carbon of various amounts of the alloying elements manganese, nickel, chromium, molybdenum and vanadium either alone or in combination. These irons remain gray. For more severe service, surface chilling is resorted to, giving a hard white layer from a thin film to sometimes several inches thick with a softer and more stress-resistant background. Castings made in this manner, including wheels, active parts of crushing machinery, metal-working rolls, etc., are produced not only of plain iron but often of alloyed iron as well.

The high silicon irons form a group of alloys that are more strictly highly-alloyed steels, as judged by composition; yet their poor mechanical properties and the processes of melting and casting generally employed permit them to be included in the discussion of the cast irons. Although with great care they may be forged, they are used only in the cast state, are difficult to cast without shrinkage defects, are non-machinable and must be finished by grinding, and are quite weak and brittle having a tensile strength of the order of 10,000 pounds per square inch and a transverse breaking strength less than half that of gray cast iron. They find no uses at elevated temperatures and although rapidly attacked by weak or strong alkalis, they are of exceptional merit in contact with inorganic acids. These alloys are much used for sulphuric and nitric acids in a wide range of temperature and concentration though suitable in contact with hydrochloric acid only when dilute and relatively cold. The compositions employed range from 13 to 17 per cent silicon, up to 1.3 per cent carbon and frequently appreciable additions of other alloying elements such as manganese, nickel

and aluminum. The most consistently good results appear however to have been obtained with silicon slightly over 14 per cent, carbon under 0.75 per cent, phosphorus about 0.15 per cent, other elements low, and with no additional alloying components.

Carbon and Low-Alloy Steels

Carbon steels, and commercially pure irons to a much lesser degree, compete with cast irons for first position in the chemical industry, in point of view of quantity used. And in other industries where corrosion at various temperatures is of great importance and contamination of the product with iron is not objectionable, simple carbon steel is still the prime material of construction. While fairly permanent in such liquids as sulphuric acid of high concentrations, mixed acid, and alkaline liquids, and also in some atmospheres, it is frequently ill suited to its environment and is therefore, except where protective coatings will adequately serve, giving way to metals that are better fitted for the severe demands and are moreover economically justified in each application. When required for permanence in the atmosphere or in either fresh or salt waters, protective coatings of either the paint, enamel, or oily types must be applied, while in such structures as boilers where carbon steels are almost exclusively employed, unless special precautions are taken, severe corrosion will ensue. Acid substances, chlorides and dissolved oxygen are often extremely damaging at modern boiler temperatures while strong hot alkaline liquids, especially in the presence of riveted construction with its consequent stress concentration, often induce serious intercrystalline attack. In the last mentioned type of service, both in boilers and in chemical engineering equipment, steels such as 5, 6 and especially 8 (of Table V) prove to be valuable and of long life.

Due to special ductility in drawing operations and a resistance to atmospheric attack appreciably better than that of ordinary sheet steel even of the better grades, a number of uses for bright sheet of commercially pure iron (No. 1 of Table V) and of electrolytic iron have developed.

Also copper-bearing soft steel has filled many needs better than ordinary carbon steel. The lower copper product (No. 16) withstands atmospheric attack, especially in industrial localities, much more effectively although failing with equal rapidity in fresh or

Table V
Carbon and Low Alloy Steels

No.	Condition	Nominal Chemical Composition C	Si	Mn	Ni	Cr	V	Cu	W	Mo	Tensile Strength 1000 Lbs. Per Sq. In.
1	Wrought	.03	.01	.02	..	..	..	..	..	..	40-50
2	Wrought	.20	.02	.35	..	..	..	..	..	..	50-65
3	Cast	.30	.30	.70	..	..	..	..	..	..	60-75
4	Wrought	.90	.25	.25	..	..	..	..	..	..	70-300*
5	Wrought	.15	.10	.50	2.0	..	..	..	..	..	60-75
6	Wrought	.15	.10	.50	3.0	..	..	..	..	..	70-90
7	Wrought	.30	.20	.70	3.5	..	..	..	..	..	70-200
8	Wrought	.10	.20	.60	5.0	..	..	..	..	..	65-140
9	Wrought	.15	.20	.30	4.0	.75	..	..	..	..	70-150
10	Wrought	.30	.20	.60	1.5	.60	..	..	..	..	70-200
10a	Cast	.30	.20	.60	1.5	.60	..	..	..	..	70-130
11	Cast	.80	.25	.75	1.0	1.00	..	..	..	..	.....
12	Wrought	.40	.20	.50	3.5	.75	..	..	..	..	100-300*
13	Cast	.50	.30	.50	2.8	.70	..	..	..	..	100-120
14	Cast	.85	.30	.75	..	1.00	..	..	..	.35	.....
15	Cast	1.70	.30	.70	.85	1.10	..	..	..	.30	.....
16	Wrought	.20	.02	.35	..	..	..	.25	..	..	50-70
17	Wrought	.25	.02	.50	..	..	..	.90	..	..	70-75
18	Wrought	.10	.15	.50	..	.60	..	.40	..	..	45-85
19	Wrought	.15	.75	.50	..	3.00	..	.50	..	..	70-150
20	Wrought	.35	.20	1.25	..	..	..	..	..	..	75-200
21	Wrought	.40	.20	1.65	..	..	..	..	..	..	85-250
22	Wrought	.85	.25	1.10	..	0.40	..	..	..	..	85-300*
23	Wrought	.20	.20	.75	..	1.00	.15	..	..	..	80-150
24	Wrought	.20	.20	.75	..	1.50	.20	..	..	..	80-170
25	Wrought	.50	.20	.65	..	1.00	.15	..	..	..	90-300*
25a	Wrought	.50	.20	.50	..	2.25	.25	..	..	..	100-300*
26	Wrought	.60	.35	.25	..	1.50	.20	..	..	..	100-300*
27	Wrought	.45	.25	.50	..	1.00	.30	..	2.5	..	100-300*
28	Wrought	.75	.25	.25	..	..	.20	..	..	..	80-300*
29	Wrought	.90	.25	.25	..	3.7	..	..	..	..	110-300*
30	Wrought	1.00	.25	.25	..	1.00	.30	..	..	..	120-300*
31	Wrought	.65	.25	.50	1.7	.70	..	..	..	.70	100-300*

*Approximate Maximum

salt waters, while the higher copper product (No. 17) has found a wide sphere of usefulness as oil lines and other piping in moist soils and under water.

Carbon steel castings of the composition shown as No. 3 are still paramount in valve bodies, fittings, etc., for temperatures up to about 660 degrees Fahr., then giving way to forged parts of the same composition or to low-alloy steels.

As service temperatures are increased, however, carbon steels become rapidly less adaptable, since much above 570 degrees Fahr. their load-carrying ability offers no inducements to the engineer. It is of course possible, and too frequently the usual procedure, when stresses at elevated temperatures are beyond the range of carbon steels or weights in the latter would not be permissible, to turn to the highly alloyed steels. Lower alloy steels, however, possess at moderately elevated temperatures (up to about 1100 degrees Fahr.)

resistance to deformation equally as good as that of many of the costlier high-alloy steels accompanied by sufficient resistance to some corrosive media to make them very desirable construction materials.

One illustration of applications of these low-alloy structural steels, has just been given in the use of nickel steel for boilers and caustic evaporators. Bolts for high temperature service in steam power plants, and in such chemical engineering work as does not involve severe corrosive attack, have been made of quenched and tempered steels of the types represented by Nos. 7, 10, 12, 20, 21, 23 and 24 of Table V. Particularly noteworthy service as to strength and general corrosion-resistance have been shown by the last-mentioned steel; on the other hand steels Nos. 20 and 21, while quite inexpensive and possessing good strength properties, are attacked by the atmosphere and by some fresh waters, though not by sea water, at about the same rate as carbon steels. The manganese steel No. 21, due to low cost and good strength is also much employed as drill pipe for deep wells. On the European continent a number of reaction chambers for ammonia synthesis and also their interior mechanism have been made of nickel-chromium steels of the type of No. 9 of Table V and of the alloys No. 15 of Table IV and No. 43 of Table VII but full satisfaction evidently does not exist for recent plant-scale experiments have been undertaken with steels of the compositions of Nos. 23 and 24 of Table V. In this country some carbon steel has been used but as lighter construction and greater surface stability accompanies the use of the alloy steels, a great number of reaction chambers and the parts functioning within them have been of steels Nos. 23 and 24 and have proven highly successful. Up to the highest temperatures used (about 1025 degrees Fahr.), load-carrying capacity has been satisfactory although in certain concentrations of the reacting gases, particularly in the vicinity of the catalyst mass, chemical combination occurs slowly and makes necessary occasional replacement of minor members. In oil refinery construction, both in the form of heavy-wall hollow-forged reaction chambers and as seamless tubing, this same material is now rendering excellent service at temperatures up to about 930 degrees Fahr.

Welded construction in these chromium-vanadium steels has been successful, sound welds having been consistently obtained with

Table VI
Proportional Limits of Steels at Elevated Temperatures

Nominal Composition	Heat Treatment (°C.)	Room Temp. Tensile Strength 1000 Lbs. Per Sq. In.	Proportional Limit At Temp. (°C.) Given Below 1000 Lbs. Per Sq. In. 20	300	400	500	600	700	Experimenter
C .25	Hot Rolled	66	20	16	13	7	.	..	French and Tucker
C .35	Normalized	93	50	..	15.5	11.5	5	..	White and Clark
C .35	Normalized	84	33	25.5	18.5	12.5	.	..	French
C .35, Ni 3.5	Normalized	105	52	35	24	13.5	.	..	French
C .10, Ni 5.0	Normalized	76.5	50	..	22	12	4	0.5	White and Clark
C .25, Ni 5.0	Normalized	104.5	30	..	22.5	11	2.5	0.5	White and Clark
C .35, Mn 1.25	Normalized	92	47.5	..	24	19	10	2.5	White and Clark
C .35, Mn 1.25	Normalized	86	44	32	21	13	..	..	French and Tucker
C .45, Mn 1.60	Normalized	100.5	67.5	..	31	22	7	3.5	White and Clark
C .40, Ni 1.5, Cr .60	Normalized	173	49*	77	46	17.5	..	..	French
C .40, Ni 1.75, Cr 1.25	Normalized	106	59	..	17.5	7	5	2.5	White and Clark
C .40, Ni 3.0, Cr 1.0	Normalized	133	51*	..	45	24.5	5	0.5	White and Clark
C .35, Cr 1.0, V .15	Normalized	144	89	74	53	22	..	..	French
C .40, Cr 2.0, V .15	Normalized	147	50	43	37	35	..	..	French and Tucker
C .40, Cr 1.0, Mo .20	820 Oil 700 Oil	...	66	38	25	13.5	5	..	French, Cross and Peterson
C .30, Cr 1.0, Mo .40	Normalized	116	36	41	36	27	..	..	French and Tucker
C .50, Cr 9, Si 3.5	Annealed	135	80	..	40	22.5	5	1.5	White and Clark
C .10, Cr 12.0	Annealed	...	36	20.5	17.5	12.0	..	..	Carpenter Steel Co.
C .10, Cr 12.0	Annealed	72	27.5	18.5	17	11	..	..	Mochel and McVetty
C .10, Cr 12.0	Quenched & Tempered	125	67	60	46.5	12	..	..	Mochel and McVetty
C .10, Cr 16.0	Annealed	...	38	..	23.5	20	12.5	4	White and Clark
C .30, Cr 20.0	Annealed	...	50	36	32	26	21.5	13.5	Carpenter Steel Co.
C .30, Cr 20, Cu 1.0	1000 Water 650 Air	...	31	22.5	10	4	1.5	..	French, Cross and Peterson
C .20, Ni 28	Normalized	77	19	13	11	8	3	..	French and Tucker
C .30, Ni 20, Cr 8	Annealed	93.5	22.5	..	19	14	11	5	White and Clark
C .25, Cr 18 Ni 23, Si 3	Annealed	...	35	27	22	17	7	..	French, Cross and Peterson
C .15, Cr 18, Ni 8	Annealed	108	18	..	14	12	7.5	6	French, Cross and Peterson
C .50, Cr 8, W 8	Annealed	...	50	..	29	19	16	7.5	White and Clark

*Low P. L. due to Air Quenching Strains

the exercise of moderate care. Whether in units on which annealing is necessitated or in those where annealing is not practiced, the resultant grain size is small and shock resistance high; further, when annealing does not follow welding, the mild air-hardening properties of this steel, in comparison with the more severe air-hardening of other alloy steels of good deformation resistance at elevated temperature, leave a desirable balance of hardness and toughness. Other manufacturing operations, such as the casting of very large forging ingots, have also proven successful in this steel.

Quite too little attention has been given to this field of "intermediate" alloy steels and new developments can be looked for here, in the near future. The data of Table VI will show the trend as also the great influence upon the load carrying ability of these

steels at moderate temperatures of the thermal treatment applied prior to use. Much promise has been shown by the chromium-copper steels such as Nos. 18 and 19 while suitable combinations of small percentages of the carbide-forming elements as in steel No. 26 may open up new uses for these moderate priced alloys.

For wearing surfaces or resistance to severe erosion many of the steels listed in Table V have been used. Among these are the gear steel types such as Nos. 12 and 25 used in the wrought form for active parts of crushers and other mining machinery and cast steels such as No. 13 along with compositions close to No. 25, used for similar purposes; steel No. 13 is also employed in railway track-work in place of Hadfield's manganese steel of 1.25 per cent carbon and 13.0 per cent manganese. Tool steels such as Nos. 4, 22, 28, 29 and 30 find their way into general industrial service wherever there exists a demand for wear and abrasion resistance and in forms other than cutting and forming tools. Steels Nos. 11, 14 and 15 are typical compositions used for rolls for continuous shaping of hot metals but would no doubt ideally fill other needs, perhaps in the mining industry. In addition to all of these, surface-hardened parts also play an important rôle. The oldest are of course, case-carburized and hardened low carbon steels such as Nos. 2, 6, 8 and 23. Recently surface hardening by combination with nitrogen formed by ammonia decomposition has attracted much attention. The surface layers obtained though thin and lacking in ductility, are extremely hard, uniformly graduated in hardness down to the softer core and quite untarnishable in air, sea water spray and many chemical agents. The process has the further advantages of requiring only a low temperature and no quenching, the result being almost total freedom from distortion in the hardening. The steels thus far used are many in number, the essential components being one or more of the carbide forming elements, either with or without aluminum. The carbon content is varied to give, in conjunction with the pre-nitriding heat treatment, the desired hardness and toughness to the metal beneath the nitrogenized layer. Applications thus far have been in service where wear is extreme and loads are neither very high nor suddenly applied.

For hot working tools requiring both strength and erosion-resistance at high temperatures, steels Nos. 29 and 31 have rendered excellent service while in the form of pneumatic tools and rock-

Table VII
Highly Alloyed Steels

No.	Condition	Nominal Chemical Composition C	Si	Mn	Ni	Cr	V	Cu	W	Mo	Co	Al	Tensile Strength 1000 Lbs. Per Sq. In.
1	Wrought	.40	.25	.50	..	4.0	..	..	6.0	..	..	..	.....
2	Wrought	.75	.35	1.00	..	4.0	..	..	6.0	..	.50	..	.....
3	Wrought	.40	.20	.20	..	5.0	.60	..	3.0	1.0	.50	..	.....
4	Wrought	.45	3.5	.60	..	8.5	..	..	..	..	..	..	90-220
5	Wrought	.50	1.0	.40	..	9.0	..	..	..	..	..	4.0	80-200
6	Wrought	.40	1.0	.50	1.5	10.0	..	..	..	..	..	2.0	90-220
7	Wrought	.50	1.0	.40	3.0	9.5	..	..	..	..	..	5.0	80-200
8	Wrought	.40	.25	.30	..	..	..	..	10.0	..	..	..	.....
9	Wrought	.45	.25	.30	..	3.0	.45	..	9.5	..	..	..	.....
10	Wrought	.65	.40	.30	..	4.0	..	..	14.0	..	..	..	.....
11	Wrought	.10	.30	.40	..	12.0	..	..	..	..	..	..	60-240
11a	Wrought	.10	1.00	.40	3.0	12.0	.20	..	3.0	..	..	..	75-260
12	Wrought	.10	.30	.40	..	13.0	..	1.0	..	..	..	..	70-240
13	Wrought	.10	.30	.40	0.6	13.5	..	..	..	..	..	.. ZrS	70-240
14	Wrought	.10	.40	.40	..	14.0	..	..	..	..	..	.40	70-210
15	Wrought	.35	.30	.40	..	13.5	..	..	..	..	..	..	80-260
16	Wrought	1.50	.50	.40	..	12.0	..	..	..	.75	3.00	..	.....
17	Wrought	1.50	.50	.40	..	12.0	1.0	..	..	1.00	..	..	.....
18	Wrought	2.20	.50	.40	..	12.0	..	..	..	..	..	..	.....
19	Wrought	.10	.60	.40	..	17.0	..	..	..	..	..	..	80-200
20	Wrought	.10	1.25	.40	..	17.0	..	..	..	..	..	..	80-150
21	Cast	.25	.75	.50	..	17.0	..	..	..	..	..	..	60-100
22	Wrought	.60	.25	.40	..	16.5	..	..	..	..	..	..	90-280
23	Wrought	.80	.20	.35	..	17.0	..	..	..	..	..	..	90-280
24	Wrought	1.10	1.00	.40	..	17.0	..	..	..	..	..	..	.....
25	Wrought	.25	.40	.50	..	20.0	..	1.0	..	..	..	..	80-125
26	Wrought	1.10	1.00	1.00	..	22.0	..	..	..	..	..	..	.....
27	Cast	.85	.50	.50	..	25.0	..	..	..	..	..	..	50-80
27a	Cast	.25	.75	.40	3.5	25.0	..	..	..	..	..	..	60-85
28	Wrought	.20	.30	.50	..	28.0	..	..	..	..	..	..	85-120
29	Wrought	.40	.25	1.50	26.0	..	..	..	..	..	..	..	75-150
30	Wrought	.20	.25	.50	33.0	..	..	..	..	..	..	..	75-150
31	Wrought	.15	.75	.60	30.0	2.0	..	..	..	..	..	..	75-150
32	Cast	3.00	1.25	.50	13.0	2.0	..	5.0	..	..	..	..	18-25
33	Wrought	.15	.30	.40	8.0	18.0	..	..	..	..	..	..	90-250
33a	Wrought	.06	.90	.40	8.0	18.0	..	..	.60	..	..	..	80-230
34	Cast	.25	.50	.40	8.0	18.0	..	..	..	..	..	..	70-90
35	Wrought	.35	1.25	.50	8.0	18.0	..	..	4.0	..	..	..	100-200
36	Wrought	.25	.30	.40	12.0	21.0	..	..	..	..	..	..	100-220
37	Wrought	.45	1.00	.75	19.5	7.5	..	..	..	..	..	..	90-200
38	Cast	.30	1.00	.75	19.5	7.5	..	..	..	..	..	..	50-70
39	Wrought	.40	1.35	.75	22.5	8.0	..	1.25	..	..	..	..	95-180
40	Wrought	.25	2.5	.50	21.0	25.5	..	..	..	..	..	..	120-200
41	Wrought	.20	2.8	.50	25.5	17.5	..	..	..	..	..	..	100-200
42	Cast	.40	1.5	.50	25.0	15.0	..	..	..	..	..	..	60-75
43	Wrought	.25	.25	1.50	32.0	12.0	..	..	4.0	..	..	..	100-200
44	Cast	.35	2.0	1.5	35.0	15.0	..	..	..	..	..	..	60-75

cutting and drilling tools Nos. 4, 27 and 28 have met with much success.

Highly-Alloyed Steels

These steels as a group are of increasing importance, yet changing constantly so that each year finds new combinations of alloying elements working upward into prominent positions in the industrial scheme and coincidently others dropping entirely out of

it. There has also been much misinformation circulated regarding their manipulation and use.

Mention has just been made of steels for hot-working operations. Among those of Table VII suited to such service are Nos. 8, 9 and 10. In past years steels No. 25, 25a and 26 of Table V were standard for dies for aluminum die castings and still are whenever other than extremely large numbers of parts of a given form are required. For extra heavy demands, however, they are being replaced by steels such as Nos. 1, 2, 3 and 9 of Table VII which are essentially hot working die steels.

Another group of highly-alloyed steels commonly used for small parts subjected to a combination of high temperature (at which they must retain both hardness and ductility) and hot gas corrosion and erosion, consists of the compositions developed for internal combustion engine exhaust valves. The low alloy steels have neither the strength nor the scaling resistance to meet the demands of modern automotive engineering, in respect to exhaust gas temperatures, velocities and compositions. Steels such as No. 10 were at one time in great favor and widely used; they retain their hardness at operating temperatures but oxidize rapidly; the plain chromium steels on the other hand, scale inappreciably but become too soft. The chromium-cobalt steel No. 16 is extremely serviceable but expensive. Hence steel No. 4 had practically established itself as the standard but is now in process of giving way, to a degree at least, to steels of the type of Nos. 5, 6 and 7. They possess slightly higher strength at elevated temperatures, equal resistance to high temperature corrosion and warp less in service due in part no doubt to the fact that the temperature cycles of use are below their thermal transformation points.

Some short-time tension test data for steel No. 4 and a number of the steels between Nos. 11 and 44 will be found in Table VI. At temperatures up to 930 and in some cases 1100 degrees Fahr., it will be noted that these alloys possess no advantage in load-carrying ability over many of the low-alloy steels. Hence, for such service their use must depend upon resistance to the attack of surrounding media. Above 1100 degrees Fahr., they will be employed for both strength and resistance to chemical combination.

Among the chromium alloy steels, the "stainless irons" or the low-carbon lower-chromium types (Nos. 11, 12, 13 and 14) are find-

ing very diversified applications. For moderate resistance to nitric acid, for exposure to atmosphere, fresh water and sea water when a thin corrosion film that does not penetrate is not objectionable, for the food products and other chemical industries, they are functioning satisfactorily. One of the principal uses has been as steam turbine blading. When the chromium content is below 13 per cent, high elastic ratios are secured by heat treatment along with high ductility and impact strength and fairly easy machining. For the same composition limits and suitable heat treatment, good resistance to corrosion exists, except perhaps in the last stages of the turbine where the steam carries suspended moisture and a peculiar pitting develops due, no doubt, to abrasion of the protective film. With this composition there results further almost complete immunity to intercrystalline weakness on silver soldering.

No. 14, a modification of the plain "stainless iron" possesses free-cutting qualities such as are found in Bessemer and open-hearth screw stocks and hence is increasing the field of corrosion-resisting products in automatic screw machine work. It further "siezes" with other metals less readily than "stainless iron" and is thus much in demand for valve spindles and the like. The presence of the finely divided zirconium sulphide does not affect the behavior toward corroding media, although it does further raise the normally high quenching temperature required for this steel.

The "stainless steel" has been the standard for immunity to the attack of fruit and vegetable juices, and other organic substances, in the manufacture of cutlery, dental and surgical instruments, etc. It has been of late, however, somewhat displaced by higher-carbon, higher-chromium alloys such as Nos. 22, 23, 24 and 26. The advantages claimed are equally as good or better resistance to staining combined with greater ability to hold a sharp edge (i. e., greater compressive strength and abrasion resistance). The higher compressive strength and resistance to wear opens for these steels a new field, in corrosion-resistant ball and roller bearings, that "stainless steel" was always quite unable to fill because of the low load-carrying ability (in this case measured as bearing endurance) of ball or roller bearings made from it.

The extremely high carbon alloys Nos. 16, 17 and 18 are of value chiefly due to extreme resistance to some forms of wear and abrasion. They do not harden to file-hardness and hence afford an

example of "soft" material being well suited to service of this character. They are used for many rubbing tools and for edged tools where the edges fail by rubbing away and will no doubt ultimately find a place in the mining industry.

To date the manufacture of nitric acid is not only taking a large portion of the production of the high-chromium low-carbon steels, such as Nos. 19 and 21 but is also responsible for the largest single installation of any of these highly-alloyed steels. This type is almost immune to all concentrations of nitric acid up to their boiling points and to the gaseous products of ammonia oxidation at temperatures up to about 1300 degrees Fahr. Entire plants have therefore been constructed of this single alloy. The cast alloy No. 21 is sensitive to variations in the casting method, yielding often large grain size and low impact strength and its manufacture should therefore be entrusted only to those skilled in its production. By raising the silicon content as in No. 20 increased resistance to oxidation is developed and tubing and sheets of this composition therefore are employed in recuperators and similar services.

Steels of the type of Nos. 25, 27 and 28 have exceptionally high resistance to those corroding media which just slightly attack the lower chromium steels, but they are used principally for their non-scaling properties at moderately elevated temperatures and their excellent behavior in sulphur containing (H_2S) hot gases. Because of the latter property they serve well in the roasting of sulphide ores and in some forms of oil still equipment. They have excellent strength at moderately elevated temperatures but above 1300 to 1475 degrees Fahr. are much inferior to the nickel-base alloys containing chromium or the high-nickel high-chromium steels. They also lack notch-toughness except under special conditions of treatment. The high carbon cast alloy No. 27 has excellent resistance to abrasion and is much used not only in structural furnace parts and containers but in members for service where this particular property is a necessity.

By applying suitable mixtures to the surfaces of molds for steel castings, these may be made of carbon or low alloy steels of any desired composition with a chromium steel surface of any desired chemical properties; such castings are now in use for furnace parts and may replace, due to obvious price advantages, some cast high-temperature appliances made entirely of these chromium steels.

They would also tend to be free of the casting defects often found in these highly alloyed steels and resulting from their high shrinkage and contraction and their sluggishness in pouring.

But the chromium steels are subject to a number of inherent defects that must be guarded against in fabrication and in use if their good qualities are to be rendered operative. Because of the air hardening properties, iron-chromium alloys containing less than 14 per cent chromium cannot be welded unless subsequent annealing is permissible. Likewise, in hot riveting, the rivets should not be heated above about 1500 degrees Fahr. for driving. This air hardening also creates manufacturing difficulties to a slight degree in the "stainless iron" but to even a greater degree in the "stainless steels." With chromium about 14 per cent and carbon low, high temperatures for hot working (over about 2000 degrees Fahr.) produce grain growth and brittleness that can be corrected only by severe reworking at correct temperature. With steels containing over 15 per cent chromium, welding produces grain growth and brittleness that cannot be corrected by annealing. For riveting, temperatures should be below 1500 degrees Fahr. especially if the silicon content is high, otherwise similar conditions result. Also in low carbon high chromium steels (16 per cent chromium or over) long duration of heating between about 840 and 1020 degrees Fahr. produces on subsequent cooling a high degree of brittleness, rendering this steel unfitted for operating temperatures within this range. The high carbon, high chromium steels air harden even more deeply than "stainless steel" and hence the same manufacturing precautions must be observed to even a greater degree. All of the high chromium steels are very rapidly attacked by sulphuric and hydrochloric acids. With 18 per cent or more of chromium they are, however, almost immune to sea water.

The high nickel steels no longer find much use as essentially corrosion-resistant materials. They appear to be very variable in character from melt to melt, some forging well and others very poorly with all conditions apparently the same and some likewise being of quite stable surface while others corrode rapidly. They do, nevertheless, find some applications, chiefly for exposure to the atmosphere and to sea water. The high carbon nickel-chromium-copper iron is a product made by the addition of monel metal and ferrochromium to low silicon pig iron and is an intermediate pro-

duct of some merit; it is useful in salt water and scales but slowly at slightly elevated temperatures (1300 to 1475 degrees Fahr.); it has, however, very low strength.

The high nickel high chromium steels form a very interesting and useful group. Their mechanical properties cannot be greatly changed by heat treatment, being the result of the degree of cold working and the manner of subsequent annealing. Corrosion re-

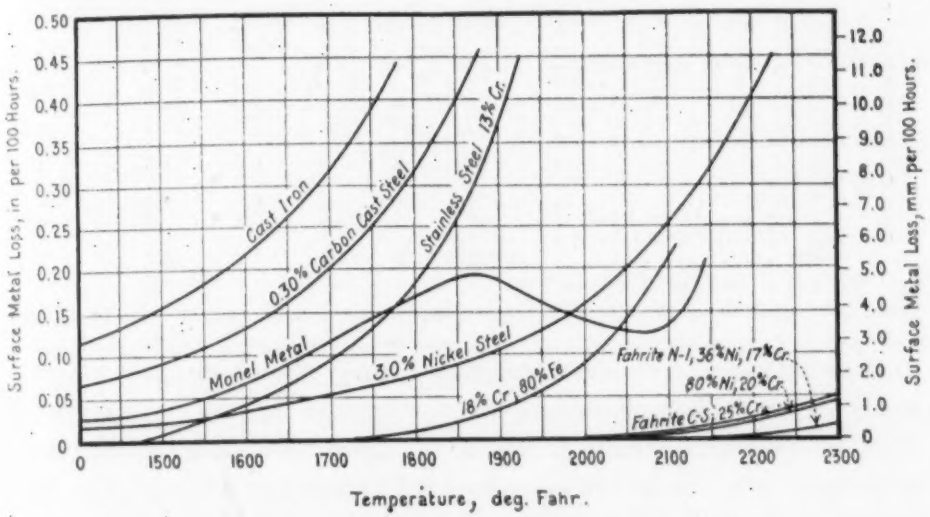


Fig. 2—Comparative Surface Loss at Elevated Temperatures.

sistance may however often be improved by quenching from high temperatures, in the vicinity of 2000 degrees Fahr. They do not function especially well in highly sulphurous hot gases, have generally low proportional limits unless severely cold-worked and are difficult to machine especially when containing over 1 per cent of silicon. The lower nickel alloys (Nos. 33, 34 and 35) serve somewhat better than the others in hot sulphide gases and are not quite as difficult to machine, and in addition are very resistant to all concentrations of nitric acid, hot and cold. The alloys may all be welded or riveted at any temperature without the production of poor mechanical properties in or near the weld or in the rivets. Those with the lower nickel content are more desirable for welding due to producing under like conditions a cleaner weld from which the welder may more readily remove oxides while working. Above 1300 degrees Fahr. the high nickeled high-chromium steels are stronger than any other of the iron-base alloys; they are therefore more difficult to hot work, being more rigid at working temperatures

Table VIII

Behavior of Some Metals in Gaseous Atmospheres at 1650 Degrees Fahr. (Hatfield)

Letter	Material	Increase in Mgs. Per Sq. Cm. for 24 Hrs. Oxygen	Steam	Carbon Di-oxide	Sul-phur Di-oxide	N_2 –72.95 O_2 – 5.0 CO_2–12.0 H_2O–10.0 SO_2– 0.05	N_2 –67.7 O_2 – 1.34 CO_2– 4.75 H_2O–21.10 CO – 5.10
A	Fe 99.9	...	...	...	....	114.0	95.3
C	Ni 99.6	...	...	...	....	5.2	16.8
G	0.17 C Steel	83.4	74.9	74.9	177.0	80.2	73.1
H	0.34 C, 3.25 Ni Steel	67.0	46.1	50.8	700.1	72.9	43.3
J	0.50 C, 3.0 Si, 8.3 Cr Steel	2.1	0.9	0.4	4.6	0.8	0.5
K	0.32 C, 1.3 Si, 13.1 Cr Steel	2.8	14.9	16.0	9.9	18.3	20.5
L	0.09 C, 18.5 Cr Steel	1.9	2.0	2.9	0.8	1.3	1.4
N	0.12 C, 17.7 Cr, 8.1 Ni Steel	2.7	6.1	3.3	1.8	2.7	0.3
O	0.35 C, 10.9 Cr, 1.4 Si, 35.2 Ni Steel	11.7	6.2	7.3	0.4	2.5	1.0
Q	0.30 C, 17.7 Cr, 7.0 Ni, 4.2 W Steel	0.2	0.4	0.3	0.5	0.4	0.2
R	11.7 Cr, 60.4 Ni, 26.2 Fe	0.8	1.4	0.5	55.6	1.4	0.8
U	70.0 Ni, 27.2 Cu, 1.7 Fe, 1.0 Mn	108.6	18.9	21.2	*	68.7	1.6

*Completely converted to compounds

and providing a narrower working range; obviously then they are more useful for severely-loaded structures at these temperatures. In high temperature strength No. 35 is stated to be of especial merit although data are lacking on its load carrying capacity. In room temperature service for corrosion resistance, they are excellent, being quite untouched by fresh and salt waters and resisting quite well both sulphuric and hydrochloric acids as well as many of the reaction mixtures and products of the chemical industries. At elevated temperatures too, they are very permanent in many active gaseous media as indicated by Table VIII and Fig. 2. As both wrought and cast products they are being rapidly introduced into oil refining equipment and, although in practically every case rendering excellent service, there has been some doubt as to the advisability of employing a material of such high initial cost for members that would greatly outlive (from the viewpoint of obsolescence) the unit of which they formed a part. On the whole, however, this group of alloys more nearly approaches the service ideal of both metallurgist and engineer, and will unquestionably expand its uses toward a constantly widening horizon.

Closure

It is apparent from the foregoing that industry's material problems are not even nearly solved—corrosion, wear and tem-

perature still account for much loss of equipment despite rapid advances in our knowledge of the phenomena and how to combat them and the concurrent strides of technology to keep abreast. For, as today's needs are approached, tomorrow's appear and keep us forever in sight of, yet never reaching, the goal. Only by rapid application of fundamental developments and constant interchange and discussion of technical applications and experiences can we hope for a closer approach.

In addition to the author's own experience with these alloys, the experiences of others have been largely drawn upon in the search for material for this paper and although no individual references appear in the text, grateful acknowledgment is here made to the authors of the publications in the following quite incomplete bibliography:

References

1. H. F. Rugan and H. C. H. Carpenter, "The Growth of Cast Irons after Repeated Heatings," *Journal*, Iron and Steel Institute, Vol. 80, 1909, page 29, Vol. 83, 1911, page 196.
2. F. Rittershausen, "Stahle fur die Chemische Industrie," Zeit. Angew. Chem. 1921, page 413.
3. M. L. Hamlin and F. M. Turner, Jr., "The Chemical Resistance of Engineering Materials," New York, (1923).
4. H. J. French and W. A. Tucker, "Strength of Steels at High Temperatures," *Iron Age*, Vol. 112, 1923, page 193.
5. Symposium on Corrosion-Resistant, Heat Resistant and Electrical Resistance Alloys, *Proceedings*, American Society for Testing Materials, Vol. 24, Part II, 1924, page 189.
6. R. S. Williams and V. O. Homerberg, "Intercrystalline Fracture in Steel," Transactions, American Society for Steel Treating, Vol. 5, 1924, page 399.
7. S. W. Parr and F. G. Straub, "The Cause and Prevention of Embrittlement of Boiler Plate," *Proceedings*, American Society for Testing Materials, Vol. 26, Part II, 1926, page 52.
8. A. E. White and C. L. Clark, "Stainless Steel," Engineering Research Bulletin No. 4, University of Michigan, Nov. 1926.
9. H. J. French, "Metals to Resist Corrosion or High Temperatures," *Transactions*, American Electrochemical Society, Vol. 50, 1926, page 47.
10. A. E. White and C. L. Clark, "Properties of Ferrous Metals at Elevated Temperatures as Determined by Short-Time Tensile and Expansion Tests," *Transactions*, American Society of Mechanical Engineers, 1927, Preprint 32.

11. D. J. McAdam, Jr., "Corrosion Fatigue of Metals as Affected by Chemical Composition, Heat Treatment and Cold Working," Transactions, American Society for Steel Treating, Vol. II, 1927, page 355.

12. S. W. Parr and F. G. Straub, "Embrittlement of Boiler Plate," *Proceedings*, American Society for Testing Materials, Vol. 27, Part II, 1927, page 52.

13. P. G. McVetty and N. L. Mochel, "The Tensile Properties of Stainless Iron and Other Alloys at Elevated Temperatures," Transactions, American Society for Steel Treating, Vol. II, 1927, page 73.

14. T. McL. Jasper, "The Building of Containers for Severe Service," *Journal of Industrial and Engineering Chemistry*, Vol. 20, 1928, page 466.

15. J. J. Kanter and L. W. Spring, "Long-Time or Flow Tests of Carbon Steels at Various Temperatures with Particular Reference to Stresses Below the Proportional Limit," *Proceedings*, American Society for Testing Materials, Vol. 28, Part II, 1928, page 80.

16. H. J. French, "Wear and Mechanical Tests of Some Railroad Bearing Bronzes," *Proceedings*, American Society for Testing Materials, Vol. 28, Part II, 1928, page 298.

17. M. G. Corson, "The Constitution of the Iron-Silicon Alloys," Iron and Steel Division, American Institute of Mining and Metallurgical Engineers, Vol. 1, 1928, page 249.

18. C. McKnight, "Recent Developments in the Use of Nickel Steels," *Transactions*, American Society of Mechanical Engineers, 1928, Preprint IS-50-6.

19. H. J. French, H. C. Cross and A. A. Peterson, "Creep in Five Steels at Different Temperatures," Technologic Paper No. 362, U. S. Bureau of Standards, Jan. 1928.

20. C. L. Clark and A. E. White, "The Stability of Metals at Elevated Temperatures," Engineering Research Bulletin No. 11, University of Michigan, Nov. 1928. Also Transactions, American Society for Steel Treating, Vol. 15, 1929, page 670.

21. B. Waeser, "Die Verwendung von Aluminium, Stahlen VA-Metallen und Eisen-Silicium-Legierungen, in der Salpetersauretechnik," *Die Chemische Fabrik*, Vol. 13, 1928, pages 529 and 544.

22. R. Furness, "Acid and Rust Proof Materials for Chemical Works," *Chemical Age*, London, 1928, page 198.

23. F. R. Palmer, "A New Development in Corrosion-Resisting Steel," Transactions, American Society for Steel Treating, Vol. 14, 1928, page 877.

THE Ac1 RANGE OF CARBON STEEL AND RELATED PHENOMENA

By Axel Hultgren

Abstract

Previous investigators have shown by other methods that the transformation of pearlite into austenite on heating requires a temperature range for its completion. The present author by a microscopic study of quenched specimens has shown the progress of this transformation. He has stated that the three factors causing this transformation to take place in a range are: (a) presence of alloy elements, (b) heterogeneity due to segregation on solidification, (c) slow diffusion rate of carbon and possibly other elements. Several conclusions and opinions are given based on observations made in this study.

Introduction

IT has been pointed out, by Kjerrman[1] and other investigators, that in ordinary carbon steels the transformation of pearlite into austenite on heating requires a temperature range for its completion, even if the rate of heating is very slow or the time at the high temperature is greatly prolonged. The method of investigation used by Kjerrman (electric resistance measurements made during the process of transformation or, at ordinary temperature, on specimens quenched from various stages of the transformation), serves well to demonstrate this fact but cannot, taken by itself, give an insight into the mechanism of the transformation.

By microscopical examination of specimens, quenched from different temperatures or after different periods of heating within the range, the progress of the transformation can be followed in detail. On the basis of such and other investigations, as well as on theoretical considerations, the following aspects of the Ac1 transformation in carbon steels may be distinguished:

1. The amount of alloying elements, such as silicon, manganese, etc., dissolved in carbon steel and as such taking part in the

[1] "The Effect of Manganese, Silicon and Phosphorus on the Pearlite Interval," Transactions, American Society for Steel Treating, Vol. IX, 1926, p. 430.

The author, Axel Hultgren, a member of the society, is metallurgical engineer, Soderfors Bruk, Soderfors Sweden. Manuscript received September 28, 1928.

transformation, raise or lower the Ac1 transformation and necessarily widen it into a temperature range.[2] Andrew and Dickie[3] have studied this phenomenon in alloy steels, where, of course, it is more pronounced than in carbon steels. For convenient reference, this temperature range will be called $Ac1_{all}$.

2. Owing to the heterogeneity of composition resulting from segregation, never completely obliterated by thermal and mechanical treatment, different transformation temperatures are required in the purer and in the segregated portions of the steel. The progress of the transformation becomes regional on a large scale and is, therefore, easily followed by microscopical examination of quenched specimens. The temperature range thus defined will be referred to as $Ac1_{seg}$.

3. It has been demonstrated by Whiteley,[4] that in commercial carbon steels of various carbon contents, when the Ac1 transformation is apparently completed, i. e., when the ferrite component of the pearlite is completely transformed, cementite particles still remain in the matrix. It will be shown that a considerably higher temperature is needed in order to bring them into complete solution. Strictly speaking, the Ac1 transformation of an eutectoid steel is not completed until this is accomplished. The range terminating in that temperature will, therefore, be designated as $Ac1_{cem}$.

A perfectly homogeneous alloy or ordinary carbon steel would show $Ac1_{all}$, but not $Ac1_{seg}$. The transformation could begin at any point of contact between ferrite and cementite. After beginning at certain points, however, it can proceed only as far as the equilibrium conditions between ferrite, cementite and austenite at that particular temperature demand, diffusion of dissolved elements taking place in the three constituents, until their changed compositions satisfy those demands, provided sufficient time be allowed. On further heating, the existing austenite nuclei will grow and possibly new nuclei form, the composition of the constituents changing again to suit the new equilibrium conditions. Thus the transformation will be regional in a sense, but the transformed regions will be more or less uniformly distributed.

[2] Kjerrman, loc. cit.

[3] "The Ac1 Range in Special Steels," *Journal*, Iron and Steel Institute, 1927, Vol. I, p. 647.

[4] "The Formation of Globular Pearlite," *Journal*, Iron and Steel Institute, Vol. I, 1922, p. 339.

In practice, where all steels are segregated to some extent, the transformation will begin in certain portions, generally the purer ones, and will end at higher temperatures in the segregated regions. In that case the regional character of the proceeding transformation is bound to the existing pattern of heterogeneity from segregation. Since all the compositions represented in the gradual transition from purer to impurer regions have their particular $Ac1_{all.}$, it is obvious that $Ac1_{all.}$, in practice, cannot generally be separated from $Ac1_{seg.}$ Since, however, the transformation assumes a regionally progressive course for both reasons, the difficulty of distinguishing those two ranges may be evaded by combining them into a common range called $Ac1_{reg.}$ When $Ac1_{reg.}$ is narrow, considering the composition, it will be due mainly to the presence of alloyed elements ($Ac1_{all.}$); when it is wide, mainly to heterogeneity resulting from segregation ($Ac1_{seg.}$).

It is convenient so to define $Ac1_{reg.}$ that its upper limit coincides with the completed transformation of the ferrite component of the pearlite, disregarding the process of solution in austenite of any cementite left in excess, whether of eutectoid or hypereutectoid origin. This definition will be adhered to in the present paper.

The Regionally Progressive Ac1 Transformation ($Ac1_{reg.}$)

In this section the results of a number of experiments will be given, in which small specimens taken from bars of different carbon steels, each specimen representing the center as well as the surface portion of the bar, were heated for some time, usually one hour, at constant temperatures within $Ac1_{reg.}$, quenched and then microscopically examined in order to ascertain the quantity and distribution of austenite formed during the heating and preserved as martensite after quenching. Details of preparation of specimens, heating and quenching are given in the appendix. Longitudinal sections of the bars were examined. The degree of transformation is defined as the area occupied by the austenite formed, in per cent of the total matrix. In each case separate figures are given for the center and the surface zone, the width of the latter being approximately one-fourth of the radius of the bar. The division into those two portions is, of course, somewhat arbitrary.

First will be presented a typical series of experiments made with one bar of 1.10 per cent carbon steel, then a series of comparative tests made with a number of bars of 1.10 per cent carbon steel of different make and finally comparative experiments showing the effect of various amounts of hot working and of annealing on the width and character of $Ac1_{reg.}$

Typical Example of Regional Transformation, 1.10 Per Cent Carbon Steel

The steel experimented upon was an annealed bar, C 3¾ inch round, made by the crucible process, of the composition given in Table I. Each specimen was held at the temperature indicated for one hour and then quenched.

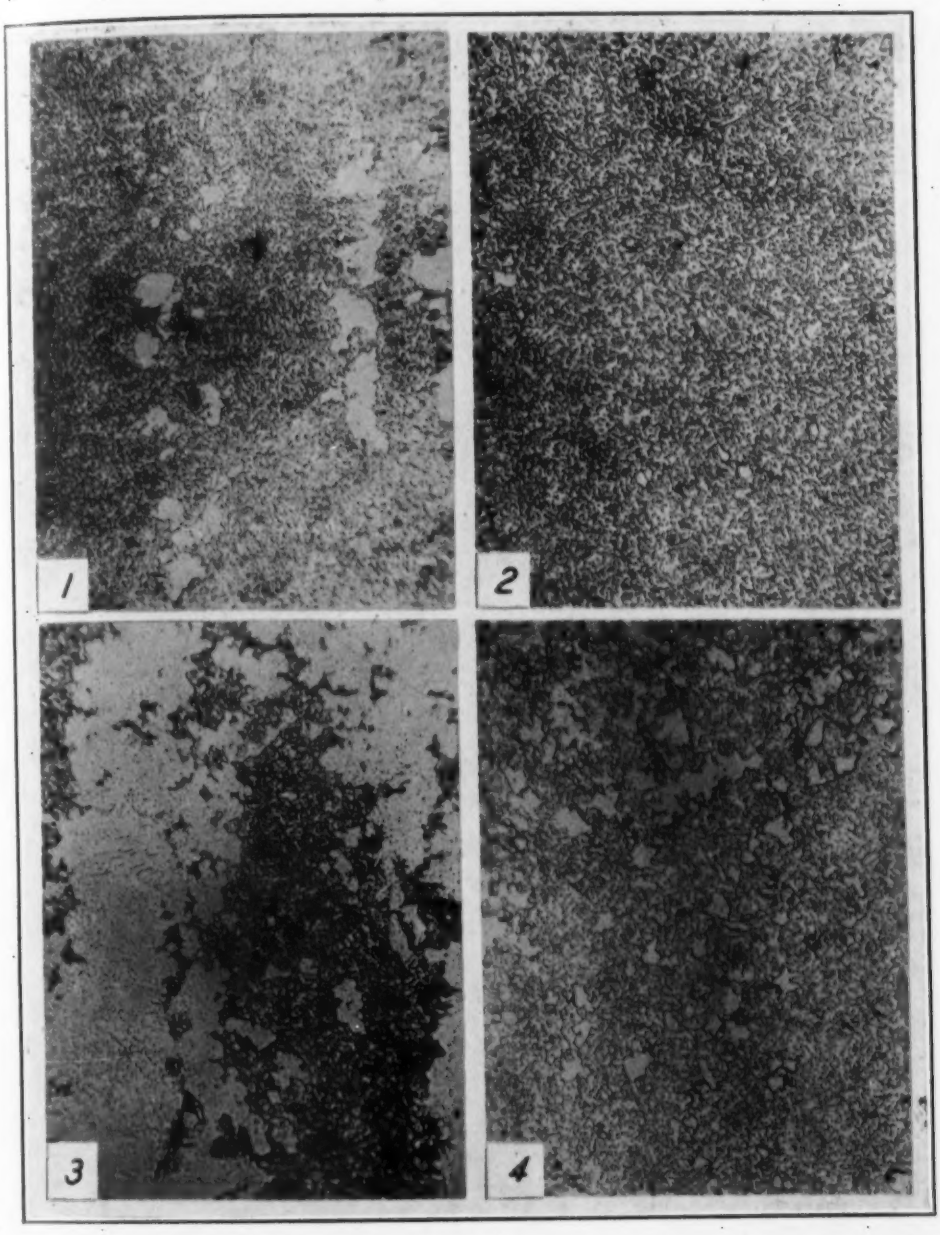
At 1337 degrees Fahr. (725 degrees Cent.) no evidence of transformation was observed. After one hour at 1342 degrees Fahr. (728 degrees Cent.) transformation had started as shown in Figs. 1 and 2, for center and surface zone respectively. In the former, rows of austenite[5] areas are seen, the inclusions present being confined to the untransformed portions. From Figs. 7 and 8, taken at high magnification, it is seen that the untransformed structure consists of globular pearlite and that the austenite areas contain free cementite. In the surface zone (Fig. 2) the austenite spots are smaller and the transformation less advanced.

The effect of heating at 1348 degrees Fahr. (731 degrees Cent.) is shown in Figs. 3 (center) and 4 (surface). In the former, the austenite now forms large continuous elongated regions, whereas in the latter the austenite areas have increased in size and number but remain disconnected.

The next pair of photomicrographs, Figs. 5 (center) and 6 (surface) represent the heating at 1353 degrees Fahr. (734 degrees Cent.). In the center, the austenite occupies the greater part of the mass, but large streaks remain untransformed. In the surface zone, however, the transformation is almost completed, only small isolated specks of pearlite being present.

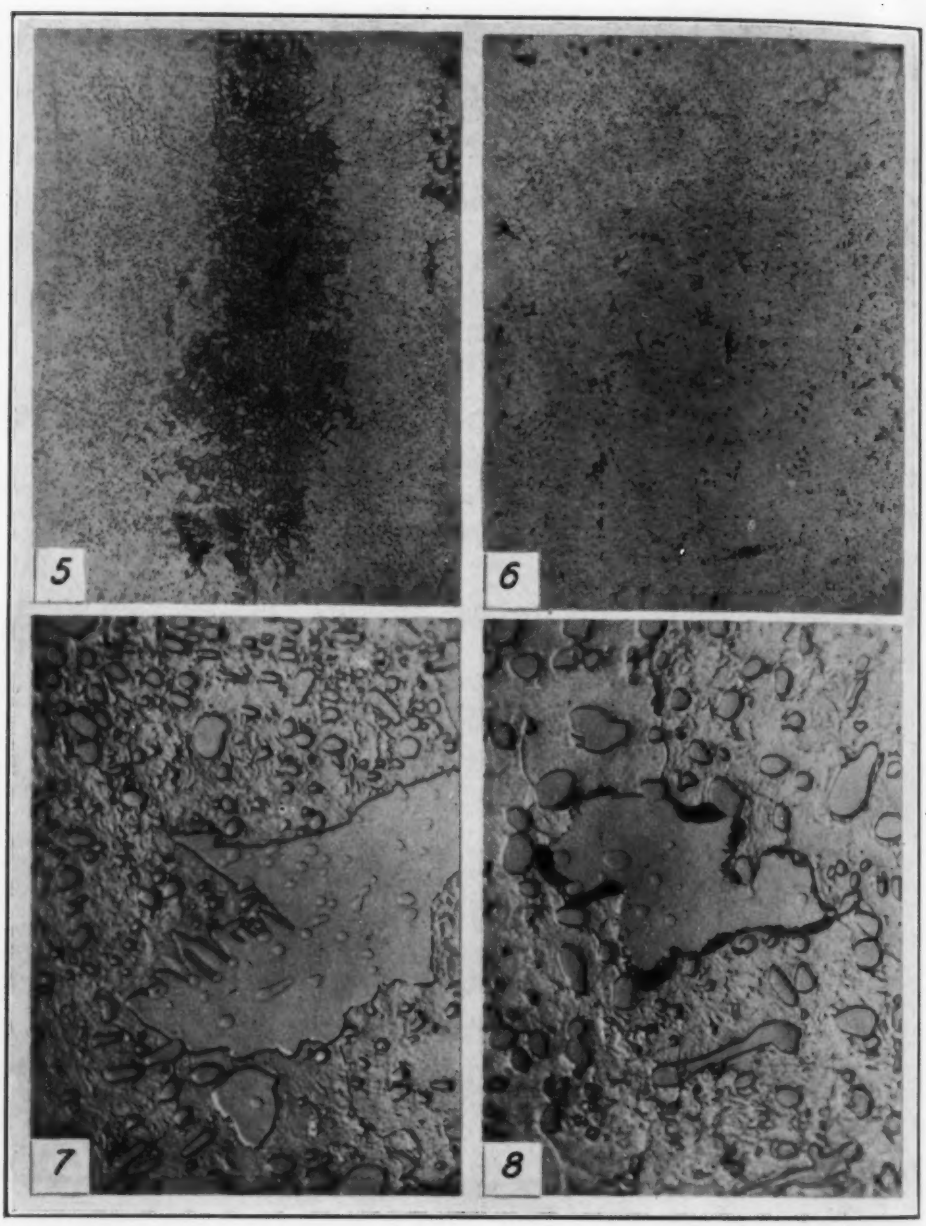
After heating at 1359 degrees Fahr. (737 degrees Cent.) the transformation near the surface was finished, whereas little progress was seen in the center after heating to 1364 degrees Fahr.

[5]For the present purpose it will be convenient to disregard the transformation into martensite on quenching.



Photomicrographs of Steel C 1.10. Samples Taken from Center and Surface of 3¾-Inch Round Bars, Quenched After 1 Hour at the Temperature Indicated. Fig. 1—1342 Degrees Fahr. (728 Degrees Cent.) Center. Fig. 2—Same as Fig. 1 but Surface Zone. Fig. 3—1348 Degrees Fahr. (731 Degrees Cent.) Center. Fig. 4—Same as Fig. 3 but Surface Zone. All Photomicrographs × 100.

(740 degrees Cent.). There was still a considerable amount of untransformed pearlite. As the temperature was raised further, the pearlite streaks shrunk more and more. Complete transformation was obtained in the center after one hour at 1429 degrees Fahr.



Photomicrographs of Steel C 1.10. Samples Taken from Center and Surface of 3¾-Inch Round Bars, Quenched After 1 Hour at the Temperatures Indicated. Fig. 5—1353 Degrees Fahr. (734 Degrees Cent.) Center. × 100. Fig. 6—Same as Fig. 5 but Surface Zone. × 100. Fig. 7—1342 Degrees Fahr. (728 Degrees Cent.) Center. × 1200. Fig. 8—Center. × 1200.

(776 degrees Cent.), but not at 1418 degrees Fahr. (770 degrees Cent.). Any sulphide inclusions present were consistently found in the streaks last to transform.

In the series of microstructures given, the course of the Ac1

transformation, as far as its regional character is concerned, is clearly depicted. There is a marked difference between the surface zone and the center. In the former, the transformation proceeds from more or less uniformly scattered nuclei and is completed in a fairly narrow temperature range, $Ac1_{reg.}$ about 1341 to 1355 degrees Fahr. (727 to 735 degrees Cent.). From those two facts it may be concluded that there is no pronounced segregation in this zone and that an appreciable part of $Ac1_{reg.}$ is due to the alloying elements present: silicon, manganese, etc. ($Ac1_{all.}$).

In the center, the transformation commences in the purer regions at about 1339 degrees Fahr. (726 degrees Cent.), consequently a little earlier than in the surface zone. With increasing temperature, it gradually spreads toward the segregated regions. When the surface zone is completely transformed [about 1359 degrees Fahr. (737 degrees Cent.)], approximately 10 per cent of the mass in the center is still untransformed. The most segregated streaks require as high temperature as about 1423 degrees Fahr. (773 degrees Cent.) for their transformation. Thus, there is marked segregation in the central part of the bar.

The difference in heterogeneity from segregation between surface zone and center portion in the present case, as indicated by the width of $Ac1_{reg.}$, 14 degrees Fahr. vs. 85 degrees Fahr., is perhaps greater than would be expected but should not surprise those who are familiar with the macrostructure of ingots as developed on sulphur prints or by Stead's and similar reagents.

Phosphorus is known to raise the alpha to gamma transformation and to be little affected by diffusion at ordinary hot working temperatures. It is, therefore, believed to be the element mainly responsible for the raised Ac1 transformation in the segregated streaks. The question, whether other segregating elements may be present, satisfying the same conditions, is outside the scope of the present paper. There is, of course, also the possibility that segregating elements exist that lower the transformation temperature, thus obscuring the effects here studied.

1.10 Per Cent Carbon Steels of Different Make

Hammered, annealed bars of 1.10 per cent carbon tool steel from five Swedish steel works were procured in the open market,

Table I
Chemical Composition and Process of Manufacture

Designation of Steel	Size of Bars in Inches	C	Si	Mn	P	S	Process of Manufacture
A 1.10	3¾	1.03	0.33	0.38	0.023	0.010	acid open-hearth
	3	1.12	0.32	0.38	0.023	0.013	acid open-hearth
B 1.10	3¾	1.09	0.06	0.37	0.024	0.007	acid open-hearth
	3	1.14	0.19	0.38	0.023	0.015	acid open-hearth
C 1.10	3¾	1.10	0.29	0.24	0.024	0.013	crucible
	3	1.18	0.14	0.32	0.023	0.010	acid open-hearth
D 1.10	3¾	1.05	0.21	0.27	0.018	0.016	acid open-hearth
	3	1.10	0.16	0.30	0.023	0.014	acid open-hearth
E 1.10	3¾	1.09	0.05	0.23	0.023	0.008	acid open-hearth
	3	1.08	0.06	0.26	0.016	0.008	acid open-hearth
Variations in composition A to E.	min.	1.03	0.05	0.23	0.016	0.007	
	max.	1.18	0.33	0.38	0.024	0.016	
F 1.10	several	1.09	0.06	0.28	0.020	0.007	acid open-hearth
F 0.80	several	0.79	0.23	0.27	0.018	0.005	basic electric

one 3¾ inches and one 3 inches round of each make. Composition and process of manufacture are given in Table I.

At each temperature, specimens from all the 10 bars A to E were treated simultaneously (for details see appendix). The temperatures employed and the results of microscopical examination are given in Table II. The percentage figures are, of course, approximate only.

From Table II the approximate temperatures at which the transformation started, $Ac1_0$, and those corresponding with 50 and 100 per cent transformation, $Ac1_{50}$ and $Ac1_{100}$, have been deduced by interpolation. These are given in Table III, arranged in the order of increasing values of $Ac1_{50}$.

From Tables II and III is seen, that for all steels examined the course of the transformation is very similar to the one described in detail for steel C 3¾ inches round in the preceding section. The following conclusions may be drawn:

1. There is little difference between the temperatures of beginning transformation, $Ac1_0$, for the steels examined: 1328 to 1344 degrees Fahr. (720 to 729 degrees Cent.). The lowest value, 1328 degrees Fahr. (720 degrees Cent.) (B 3¾ inches) may be traced to the combined effect of low silicon and high manganese content.

2. In the first stage, up to roughly 50 per cent, there is a tendency for the transformation to advance further in the center than in the surface zone. This is possibly due to "negative" segregation in the center.

3. The temperature of 50 per cent transformation, $Ac1_{50}$,

Table II

Degree of Transformation, at Different Temperatures, of 1.10 Per Cent Carbon Steels of Different Make; Center (C) and Surface Zone (S)

Size of Bars in Inches	Designation	°C / °F	722 / 1332		Degree of Transformation, Per Cent, after One Hour at																							
					725 / 1337		728 / 1342		731 / 1348		734 / 1353		737 / 1359		740 / 1364	743 / 1369	746 / 1375	749 / 1380	752 / 1386	755 / 1391	758 / 1396	764 / 1407	770 / 1418	776 / 1429			
			C	S	C	S	C	S	C	S	C	S	C	S	C	C	C	C	C	C	C	C	C	C			
3¾	A		..	..	0	0	<1	<1	50	6	75	90	95	100	98	99	>99	>99	100	..	..	..	..	..			
	B	15	10	60	60	60	95	99	99	100	99	..	99	..	99	>99	>>99	>>99	>>99	..	..	..	..	..			
	C	..	..	0	0	0	10	1	25	10	80	98	95	100	98	99	>>99	>>99	>>99	>99	>99	>99	>99	100			
	D	..	..	0	..	..	<1	0	75	60	95	100	98	..	99	99	>>99	>>99	100	..	..	..	..	..			
	E	0	..	1	0	0	90	97	95	>99	97	100	99	..	99	99	>>99	>>99	>>99	>99	100	..	..	..			
3	A	..	..	0	0	0	4	1	50	25	75	90	97	100	99	99	>>99	>>99	100	..	..	..	..	..			
	B	..	..	0	0	0	25	20	65	75	90	99	98	100	99	99	>>99	>>99	>99	100	..	..	..	..			
	C	..	..	0	0	0	40	40	85	90	97	100	99	..	99	>99	>>99	>>99	100	..	..	..	..	..			
	D	..	..	0	0	0	5	2	75	90	97	>99	99	100	99	>>99	>>99	>>99	>99	100	..	..	..	..			
	E	0	0	<1	<1	90	99	98	100	99	..	..	99	..	>99	100	..	..	..	..	..	..	..	..			

is closely similar for center and surface zone of each steel and seems to be a function of the composition. The effect of silicon to raise and manganese to lower the alpha to gamma transformation is reflected in the parallelism in the variation of $Ac1_{50}$ and the term (2 Si–Mn), as seen from Table III. Without claiming any high accuracy, it may be deduced from the present results, that 0.10 per cent silicon raises the transformation by 5.4 degrees Fahr.

Table III
Condensed Transformation Data for 1.10 Per Cent Carbon Steels

Designation and Diameter of Bars in Inches	2 Si–Mn per cent	$Ac1_0$ °C		$Ac1_{50}$ °C		$Ac1_{100}$ °C		Width of $Ac1_{reg}$ °C	
		C	S	C	S	C	S	C	S
B 3¾	—0.25	720	721	724	724.5	750	729	30	8
E 3	—0.14	724	724	726.5	726.5	741	729	17	5
E 3¾	—0.13	724	726	726.5	726.5	761	732	37	6
C 3	—0.04	726	726	728.5	728.5	747	733	21	7
B 3	+0	726	726	729.5	729.5	753	735	27	9
D 3	+0.02	726	727	730	729.5	753	735	27	8
D 3¾	+0.15	727	729	730	730.5	753	733	26	4
A 3	+0.26	726	727	731	732	747	736	21	9
A 3¾	+0.28	727	727	731	733	750	736	23	9
C 3¾	+0.34	726	727	732.5	732.5	773	735	47	8

(3 degrees Cent.), whereas 0.10 per cent manganese lowers it by 2.7 degrees Fahr. (1.5 degrees Cent.).

4. As the transformation passes about 50 per cent, a marked difference between surface zone and center becomes apparent. In the former it is rapidly completed, the total range $Ac1_{reg.}$ for the steels examined varying between 9 and 16 degrees Fahr. (5 and 9 degrees Cent,). From results given later in Table IV it may be inferred that those values of $Ac1_{reg.}$, however small as compared with the values obtained in the center, are greater than would be accounted for merely by the presence of alloying elements ($Ac1_{all.}$). Consequently, there is a certain amount of segregation in the surface zone also.

5. The center, in all steels examined, exhibits a marked heterogeneity from segregation, the $Ac1_{reg.}$ varying between 38 and 85 degrees Fahr. (21 and 47 degrees Cent.). It may be seen, however, that a degree of transformation of 95 to 99 per cent is reached at moderate temperatures, whereas the last few per cent of the mass require considerably higher temperatures. This indicates, that the bulk of the segregated elements is confined to narrow streaks in the bar.

The mechanism of the regional transformation is further illustrated by the photomicrographs taken from the present series of tests with 1.10 per cent carbon steels.

Fig. 7 shows the appearance at high magnification of a small area of austenite in the process of growth in globular pearlite. On detailed examination one may see, advancing into the ferrite, several parallel branches, separated by cementite particles at their bases. It may be inferred that there is a favored direction of growth of the austenite, probably governed by the crystal orientation of the ferrite. This would explain why the austenite did not first grow around the cementite particles, from which the carbon necessary for the transformation is supplied. Fig. 8 shows troostite fringes at the circumference of an austenite area. The conditions of their formation in this case would seem to be inefficient cooling in quenching aided by inoculation from the adjacent ferrite.

Figs. 9 and 10 are taken from different portions of a segregated streak in the process of transformation. Both show systems of parallel branches of advancing austenite. In the former, two sets of parallel branches are seen that apparently grow from opposite sides of the ferrite area, thus the assumption being strengthened that the orientation of the ferrite crystal determines the favored direction of austenite growth. In Fig. 10 two sets of austenite branches are seen to intersect, an indication that one ferrite grain presents more than one favored direction of austenite growth. This would seem plausible.

Effect of Hot Working and Annealing on $Ac1_{reg.}$

In this series two steels, F 1.10 and F 0.80, were investigated. The compositions are given in Table I. The former was an acid open-hearth steel with 1.09 per cent carbon and 0.06 per cent silicon; the latter was made in the basic electric furnace; it contained 0.79 per cent carbon and 0.23 per cent silicon. In both cases 9-inch square ingots were forged to different sizes ranging from 4 inches to ¼ inch rounds. All sizes of each steel were taken from the same portion of one ingot. Specimens of unannealed as well as annealed steels were tested. All results of this series are brought together in Table IV.

If the unannealed specimens are compared with the annealed ones taken from the same size of bar, it is seen, for both F 1.10

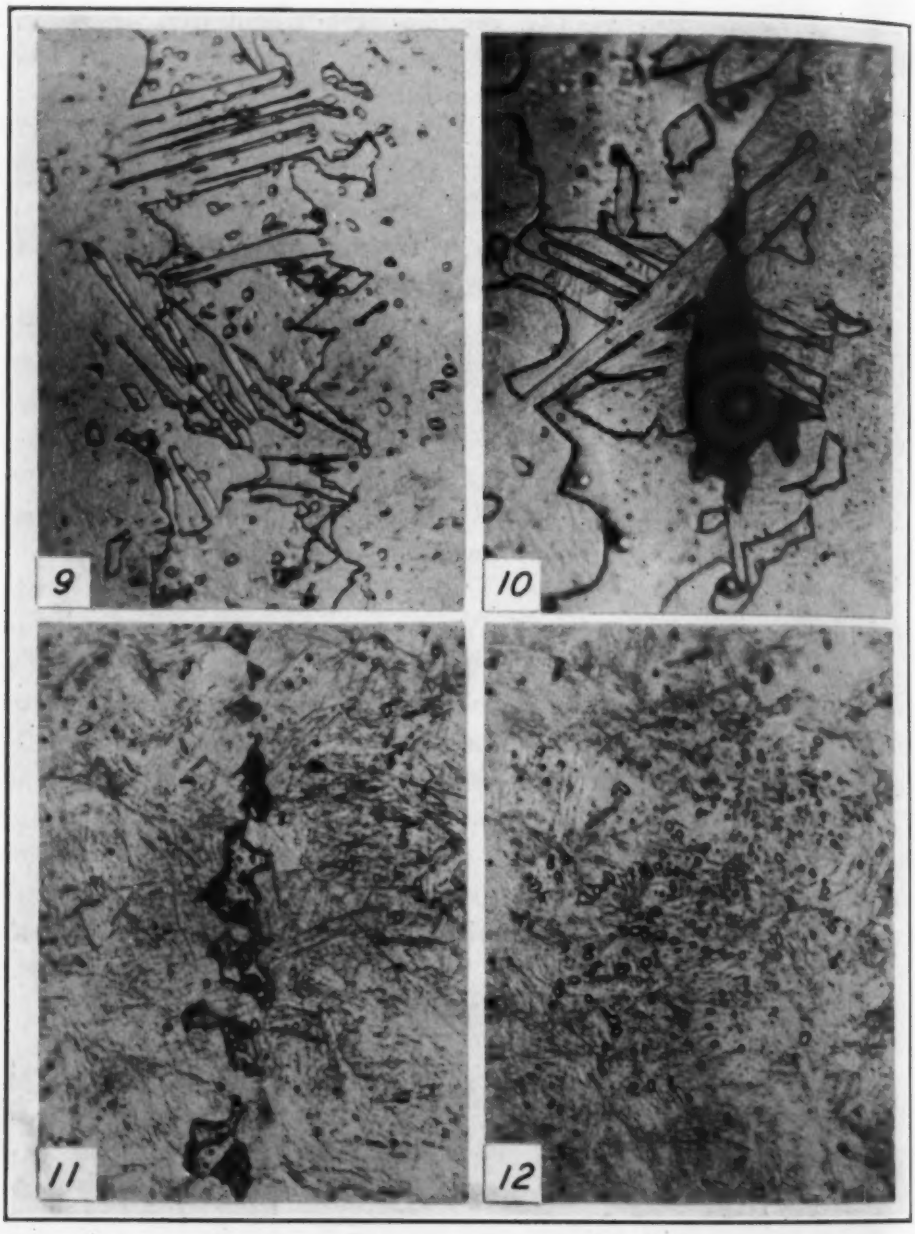


Fig. 9—Photomicrograph of Steel E 1.10, Sample Taken from Center of 3-Inch Round Bar Quenched After 1 Hour at 1375 Degrees Fahr. (746 Degrees Cent.). × 1200. Fig. 10—Same as Fig. 9. Fig. 11—Photomicrograph of Steel F 0.80, Sample Taken from Center of ½-Inch Round, Unannealed Bar, After 1 Hour at 1364 Degrees Fahr. (740 Degrees Cent.). × 1200. Fig. 12—Same as Fig. 11.

and F 0.80, that for all temperatures within the main part of the transformation range, the degree of transformation is greater for the unannealed steels than for the corresponding annealed steels. They coincide above 99 per cent transformation. This applies to

Table IV

F 1.10 and F 0.80 Unannealed and Annealed Steel, with Various Degrees of Hot-Working

U = Unannealed A = Annealed C = Center S = Surface Zone

Degree of Transformation, Per Cent after One Hour at

Steel	Diameter of Round Bars in Inches and Condition	°C / °F	722 / 1332		725 / 1337		728 / 1342		731 / 1348		734 / 1353		737 / 1359		740 / 1364		743 / 1363		746 / 1375		752 / 1386		758 / 1396		764 / 1407		770 / 1418
			C	S	C	S	C	S	C	S	C	S	C	S	C	S	C	S	C	S	C	S	C	S	C	S	C
F 1.10	4 A	..	0	0	15	5	97	100	98	..	99	..	99	..	>99	..	>99	..	>99	..	>99	..	>99	..	100	..	..
	2 A	..	0	0	20	2	97	>99	98	100	99	..	99	..	>99	..	>99	..	>99	..	>99	..	..	..	>99	100	100
	2 U	..	1	<1	65	30	98	>99	99	100	99	..	99	..	>99	..	>99	..	>99	..	>99	..	..	..	100	..	..
	1 A	..	0	0	10	9	99	100	>99	..	100	..	..	..	..	..	..	..	..	..	..	..	..	..	..	..	..
	½ A	..	0	0	8	15	>99	100	100	..	..	..	..	..	..	..	..	..	..	..	..	..	..	..	..	..	..
	¼ A	..	0	0	4	10	100	100	..	..	..	..	..	..	..	..	..	..	..	..	..	..	..	..	..	..	..
F 0.80	2 A	..	0	0	0	0	20	13	50	50	98	99	99	>99	>99	100	>99	..	>99	..	100	..	..	..	..	..	..
	2 U	..	4	1	20	7	55	90	85	>99	99	>99	99	100	>99	..	>99	..	>99	..	>99	..	100	..	..	..	..
	½ U	..	2	1	15	25	70	80	90	>99	99	100	>99	..	>99	..	100	..	..	..	..	..	..	..	..	..	..

center as well as surface portion and will be further discussed. The temperature of completed transformation, $Ac1_{100}$, seems to be independent of whether the steel is annealed or not. Since $Ac1_0$ is lower for the unannealed steel, it follows that $Ac1_{reg.}$ is wider. In consequence, when two steels are compared in regard to heterogeneity from segregation, on the basis of $Ac1_{reg.}$, both steels should preferably be in the same condition of annealing, or else the lower portion of the transformation curve should be disregarded.

The degree of reduction in hot working is reflected in the position of $Ac1_{100}$ for the series of sizes investigated of steel F 1.10. The effect is most plainly seen in the center.

The values of $Ac1_{reg.}$ in the center for the various sizes of F 1.10, annealed, are approximately:[6]

2 and 4-inch rounds	67 to 77 degrees Fahr. (37 to 43 degrees Cent.)
1-inch round	12.6 degrees Fahr. (7 degrees Cent.)
½-inch round	8.1 degrees Fahr. (4.5 degrees Cent.)
¼-inch round	5.4 degrees Fahr. (3 degrees Cent.)

Thus, when the size is reduced below 2-inch rounds, i. e., at a reduction from the size of the ingot exceeding 25 to 1, $Ac1_{100}$ is lowered, the more the greater the amount of reduction, while $Ac1_0$ is slightly raised. It should be added that tests made on specimens, taken from ingots of similar steel but not from the same heat, agree well with the present results for the 4-inch round bar.

It may be inferred that, when a certain degree of reduction is reached, the steel is homogenized by diffusion at the high temperature of heating and working. This question has not been pursued further in the present investigation, but it seems probable that a study of the effect of temperature, time of heating and size of billet employed in relation to size of ingot, according to the present method, would give results of theoretical and practical interest. $Ac1_{reg.}$ for the surface zone is very small for steel F 1.10, but large for steel F 0.80. There are reasons to believe that the latter fact is due to the presence of blowhole segregation in this zone.

Effect of Duration of Heating on Degree of Transformation

Series of specimens of the two steels F 1.10 and F 0.80 were held for different periods, from 5 minutes to 3 hours, at 1342

[6]The approximations are admittedly somewhat crude for the narrower ranges, owing to: 1. unavoidable temperature fluctuations during the tests, and 2. the temperature increments of 5.4 degrees Fahr. (3 degrees Cent.) employed.

Table V

Different Periods at 1342 Degrees Fahr. (728 Degrees Cent.)

U = Unannealed A = Annealed C = Center S = Surface Zone

Steel	Diameter of Round Bars and Condition	Degree of Transformation, Per Cent, at 1342 Degrees Fahr. (728 Degrees Cent.) after 5 minutes C	5 minutes S	10 minutes C	10 minutes S	20 minutes C	20 minutes S	1 hour C	1 hour S	3 hours C	3 hours S
	4 A	40	30	75	60	90	95	97	100	98	..
	2 A	40	30	50	x	65	90	97	100	98	..
F 1.10	2 U	75	45	90	99	95	100	98	...	99	..
	½ A	30	50	90	98	97	99	100	100	..	..
	½ U	80	95	99	100	99	..	>99	...	100	..
	½ A	10	2	20	20	30	20	35	25	55	35
F 0.80	½ U	40	40	70	80	75	80	80	98	90	99

x Not determined.

degrees Fahr. (728 degrees Cent.). The results are given in Table V.

The results of this series of experiments may be summed up as follows:

1. The Ac1 transformation of a given steel at constant temperature within $Ac1_{reg.}$ proceeds faster in the unannealed than in the annealed condition.
2. In annealed bars, the transformation proceeds faster, the more advanced the stage of reduction in hot working.
3. The degree of transformation obtained after one hour appears not to be far below an assumed asymptotic value for the temperature in question. Thus one hour may be considered sufficient as a heating period in tests for comparative purpose.

Retarded Solution of Pearlitic Cementite in Austenite, $Ac1_{cem.}$

As already pointed out, the Ac1 transformation of eutectoid pearlite, in carbon steels, is not completed on the disappearance of the pearlitic areas in the structure, since part of the cementite, at that stage, remains undissolved. In other words, the eutectoid pearlite behaves as if it had hypereutectoid composition.

For the study of this phenomenon steel F 0.80 (Table I) with 0.79 per cent carbon was selected. Fig. 11 shows the structure in the center of a ½-inch round specimen, unannealed, heated for one hour at 1364 degrees Fahr. (740 degrees Cent.) and quenched. A row of untransformed pearlite[7] areas are seen. The

[7] No free ferrite could be distinguished in the untreated specimen. Obviously, the cooling of this small dimension during or after forging had been so rapid as to suppress the separation of the proeutectoid constituent. The pearlite should, therefore, be slightly hypoeutectoid.

main mass, being transformed in the sense that it is free from pearlite, contains scattered cementite particles. Fig. 12 shows another field of the same specimen, where the number of undissolved cementite particles is particularly prominent.

After one hour at 1369 degrees Fahr. (743 degrees Cent.) no pearlite remained, in other words $Ac1_{reg.}$ was completed, but the quantity of free cementite was not sensibly diminished. It follows, that at this stage the austenitic matrix contained, on an average, less than 0.79 per cent carbon. In a series of specimens heated for one hour at increasing temperatures the cementite gradually diminished, but the last trace was not dissolved until 1418 degrees Fahr. (770 degrees Cent.) was reached.

Thus, for a heating period of one hour, the Ac1 ranges for steel F 0.80, ½-inch round, unannealed, center portion, were:

$Ac1_{reg.}$ = 1330 to 1366 degrees Fahr. (721 to 741 degrees Cent.)
$Ac1_{cem.}$ = 1330 to 1418 degrees Fahr. (721 to 770 degrees Cent.)

Distribution of Dissolved Carbon in Austenite, Slow Diffusion in Austenite

In this microscopic examination of hardened steels, in which the structure before hardening consisted of lamellar pearlite, faint lamellar markings had been repeatedly observed. Accidentally, an etching reagent was found that clearly brought out this feature.

The reagent, as finally employed, was prepared by dissolving 2 grams of picric acid and 25 grams of sodium hydroxide in 50 cubic centimeters of water. By immersion of a hardened specimen for 1 to 3 minutes in the boiling solution, the polished surface assumes a yellowish brown tint, which is the correct one for the present purpose. If, in that time, it turns dark violet, the solution is too strong and should be diluted with water. If no staining is obtained, some water should be boiled off. The reagent will be referred to as the concentrated sodium picrate reagent.

A portion of the ½-inch round hammered bar of steel F 0.80 (Table I) was heated to 1472 degrees Fahr. (800 degrees Cent.) and allowed to cool in the furnace. The structure obtained consisted almost entirely of lamellar pearlite. From this bar a series of half-cylindrical specimens were prepared, which were then heated to various temperatures, by immersion in a salt bath, and quenched in water. The flat surface, representing a longitudinal

section of the bar, was polished and etched in the concentrated sodium picrate reagent.

Fig. 13 shows the structure of a specimen held for 10 minutes[8] at 1359 degrees Fahr. (737 degrees Cent.) and quenched. The matrix is, of course, finely martensitic, but no needle structure is to be seen after the etching employed. Instead, a system of light lamellar markings appear, occupying the position of the cementite lamellæ of the pearlite, that existed prior to the transformation into austenite. In some of the light lamellæ dark dots, representing undissolved cementite, are seen. Some cementite particles were found which were oblong in the direction of the lamellæ; as a rule, however, they were found to have a rounded shape.

A specimen quenched after 5 minutes at 1418 degrees Fahr. (770 degrees Cent.) showed no undissolved cementite remaining, but the lamellar pattern was still very marked (Fig. 14). Fig. 15 shows the structure of a specimen quenched after one hour at 1418 degrees Fahr. (770 degrees Cent.). The lamellæ are now somewhat fainter. Furthermore, a network, indicating the grain boundaries of the austenite, is seen. Those boundaries pass indiscriminately across the lamellæ.

On heating at still higher temperatures, the lamellæ of the quenching structure became less marked, the more so as the time of heating was prolonged. This may be seen from the following data:

1454 degrees Fahr. (790 degrees Cent.), 30 minutes	lamellæ fairly well visible
1454 degrees Fahr. (790 degrees Cent.), 1 hour	presence of lamellar markings uncertain
1472 degrees Fahr. (800 degrees Cent.), 30 minutes	faint lamellæ
1472 degrees Fahr. (800 degrees Cent.), 1 hour	uncertain

As the time at these temperatures increased, the needle structure of the martensite became more and more evident. This tended to mask the lamellar pattern. Obviously, as the temperature was raised, shorter periods of heating were required in order to efface the lamellar structure. In all specimens a segregation pattern was in evidence, the purer regions assuming a lighter stain than the segregated streaks (see Fig. 15). This effect of sodium picrate etching was first pointed out by E. Walldow (unpublished investigation, 1927).

From the series of structures shown it may be concluded, that

[8] The time given, in each case, is the total time of immersion.

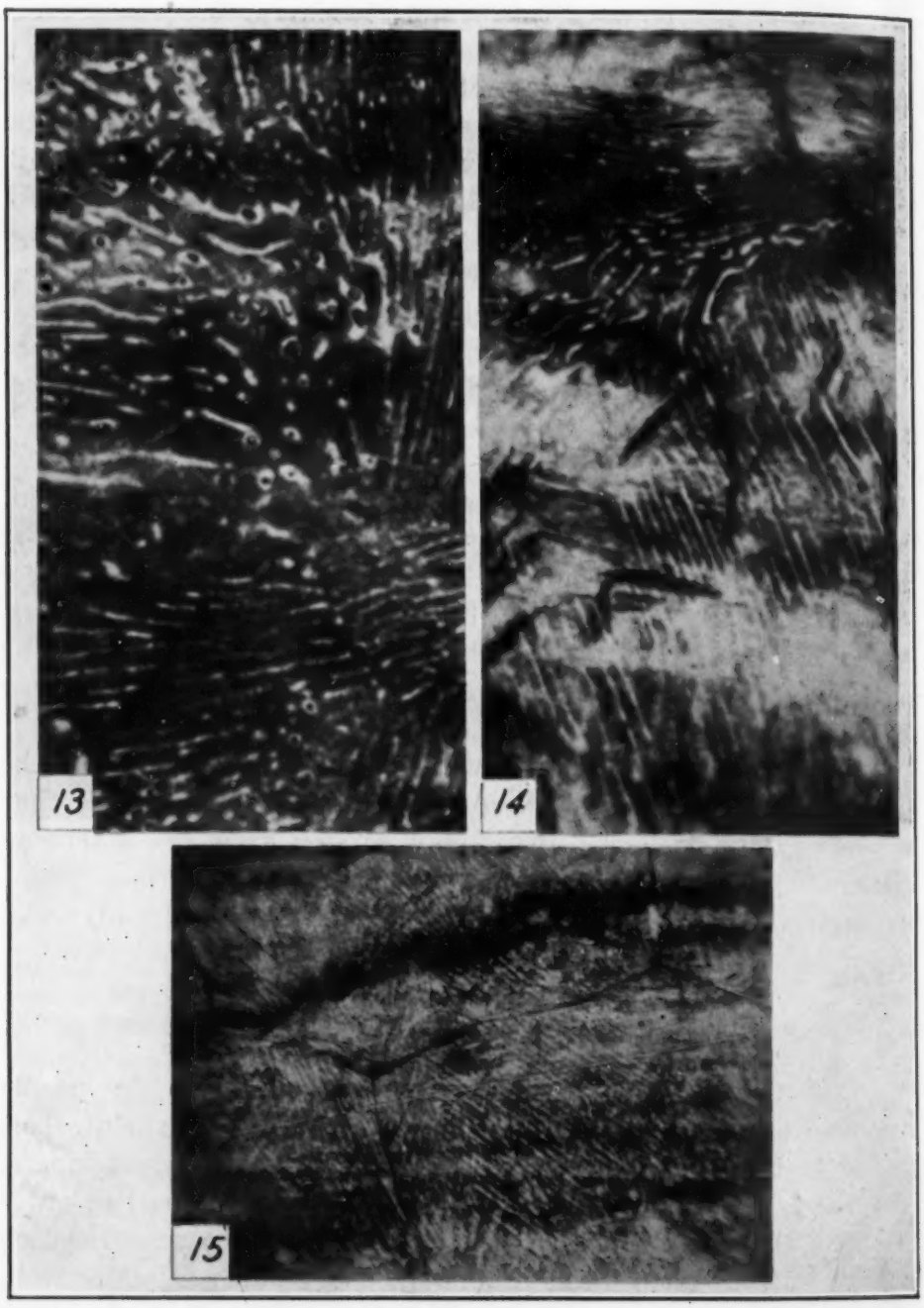


Fig. 13—Photomicrograph of Steel F 0.80, ½-Inch Round, Quenched After 10 Minutes at 1359 Degrees Fahr. (737 Degrees Cent.). × 2400. Fig. 14—Steel F 0.80, ½-Inch Round, Quenched After 5 Minutes at 1418 Degrees Fahr. (770 Degrees Cent.). × 2400. Fig. 15—Steel F 0.80, ½-Inch Round, Quenched After 1 Hour at 1418 Degrees Fahr. (770 Degrees Cent.). × 1200. All Specimens Etched with Concentrated Sodium Picrate.

the austenite formed in the Ac1 transformation is of heterogeneous composition, the pattern of heterogeneity closely conforming with

the structure of the pre-existing pearlite. Presumably, the cementite lamellæ or globules of the pearlite, on dissolving, give rise to portions in the austenite, which are richer in carbon than those taking the place of the ferritic portions. On heating to increasing temperatures, this heterogeneity is gradually eliminated by diffusion, but, in the case examined, heating for ½ hour at 1472 degrees Fahr. (800 degrees Cent.) was not sufficient for this purpose.

In quenched steels, comparatively coarse needles of troostite may be sometimes found near the surface. It is believed that those needles are martensite formed during an arrest in the cooling, coinciding with the temperature of the austenite-martensite transformation and thus favoring the development of large needles, which are then quickly tempered into troostite owing to the comparatively high temperature. A local steam cushion formed in the early part of the quenching operation, while the temperature is high enough for the Leidenfrost phenomenon to be maintained, would produce this effect. Under such conditions, it is even likely that, owing to the insulation offered by the steam cushion, heat flow from within will, for an instant, reheat the layer in which large martensite needles have appeared and thus effectively temper them. In the next moment cooling will be resumed and the untransformed austenite will be transformed into fine martensite.

Such surface troostite may be seen in Fig. 14, which shows a portion of the specimen quenched after 5 minutes at 1418 degrees Fahr. (770 degrees Cent.). The large troostite needles are stained darker than the martensitic matrix. It is of interest to note, that, while the shape of some of the needles is apparently undisturbed by the lamellar heterogeneity of the matrix, others seem to be somewhat deflected or restricted in their growth by the high carbon lamellæ. It is also noteworthy that the lamellar heterogeneity is visible within some of the needles.

This condition of the austenite, revealed by the concentrated sodium picrate reagent as applied to hardened carbon steel, may be termed "pearlitic" heterogeneity, since it is inherited from the distribution pattern of the constituents of the pre-existing pearlite. The fact that it persists up to 1472 degrees Fahr. (800 degrees Cent.) and probably higher temperatures, depending upon the time element, shows that the rate of diffusion of carbon, and probably of other elements present, is slow in the temperature range in question, at least for a certain range of carbon contents, within

the limits set by the constitution diagram. This will be discussed presently.

The existence of a wide temperature range of slow diffusion of carbon and other elements in the austenite, and the resulting heterogeneity, within this range, of all austenite formed from pearlite, i. e., in heating, would appear to constitute two facts of fundamental importance in connection with several phenomena observed in carbon as well as alloy steels, including the main theme of the present paper, the Ac1 range. It is to be expected that factors tending to retard diffusion, such as low temperature and the presence of certain dissolved elements, will render such effects particularly conspicuous. Accordingly, an attempt will be made briefly to discuss some well-known phenomena in the light of the two facts emphasized, beginning with the Ac1 transformation.

Phenomena Related to the Existence of a Range of Slow Diffusion of Carbon and Other Elements in Austenite

1. The Ac1 Range as Affected by Retarded Solution of Cementite in Austenite

Fig. 16 shows in a schematic manner a portion of the iron-carbon diagram. A_1 is the eutectoid line. It is assumed that, on heating a pure eutectoid steel, no austenite will form, until the temperature denoted by the line BC is reached. To the right in Fig. 16 is shown an austenite grain formed at the interface between a cementite grain and the adjacent ferrite. The processes involved in the growth of the austenite, i. e., in the progress of the Ac1 transformation, are:

a. "solution" of cementite at the surface c, this surface migrating further into the cementite. The mechanism of this process may be described as: decomposition of the cementite lattice, the iron atoms joining the austenite lattice and the carbon atoms entering into the interstices. Owing to the high carbon content of the cementite, the rate of solution of cementite is governed by the velocity with which the surplus carbon atoms are carried away by:

b. diffusion of carbon within the austenite from surface c to surface b,

c. "solution" of ferrite at surface b, this surface advancing

into the ferrite; in other words, transformation of body-centered ferrite near b into face-centered austenite, as the carbon required for this transformation is supplied by diffusion from c.

If we assume that equilibrium is constantly being maintained at the boundaries b and c the carbon contents of these boundaries will be B and C respectively and the growth of the austenite grain

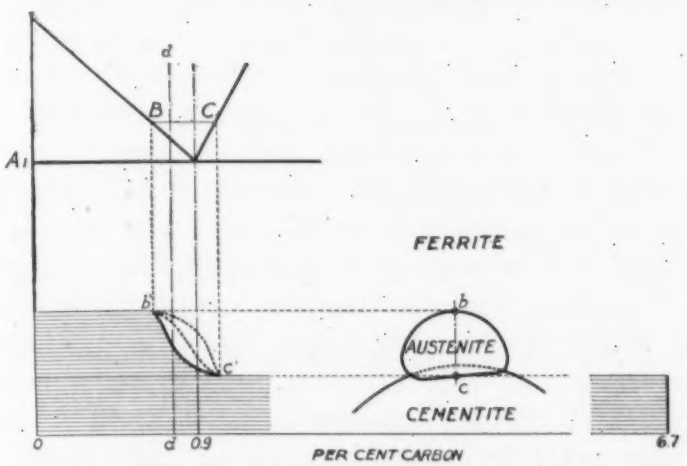


Fig. 16—Schematic Diagram Illustrating Carbon Distribution in a Grain of Austenite Growing in Pearlite.

will be governed by the rate of diffusion of carbon only. If the diffusion were rapid, one of the dotted curves b′-c′ shown to the left in the figure would represent the carbon distribution in the austenite during this process. Since, however, it has been established that the rate of diffusion of carbon in austenite is very slow, even up to 1472 degrees Fahr. (800 degrees Cent.), and must be still slower at lower temperatures, the carbon distribution is more correctly represented by the full-drawn curve b′-c′.

We assume the case that the pearlite is built up of parallel lamellæ; further, that the direction of diffusion is perpendicular to the lamellæ. In that case the average carbon content of the austenite would be as shown by the line d-d, consequently less than the eutectoid content 0.9 per cent. Thus the fact is explained, that in the Ac1 transformation of eutectoid pearlite, when all the ferrite is transformed, there is still some cementite left undissolved.

If the pearlite is globular, the diffusion paths, in the earlier stages of the transformation, will probably diverge from the cementite particles into increasing areas, thus shifting the average

carbon content d-d still further to the left and retarding the transformation. As the austenite grows to include several cementite particles, this effect will gradually recede.

The coarser the cementite lamellæ or globules the longer will be the distance for carbon to diffuse and the slower, in consequence, the transformation.[9] That is probably the main reason why, in unannealed steels, disregarding the segregated portions, the Ac1 regional transformation proceeds faster and is completed at lower temperatures than in the corresponding annealed steels. The explanation advanced by Kjerrman,[10] who concludes that heterogeneity in the distribution of alloyed elements, caused by quick cooling through Ar1, is responsible, may be partly true, but the objection may be raised that any lowering of Ac1 in certain areas, caused by higher manganese, would have as a necessary complement a raised Ac1 in areas impoverished in that element.

The fact that the austenite may grow in certain favored directions in the ferrite, rather than spread along the cementite-ferrite interface (see Fig. 7), may be interpreted as indicating that the rate of carbon diffusion is more rapid at lower than higher carbon contents, at least between the limits B and C. Hereby the average carbon content of the austenite formed is still further reduced and the quantity of cementite left in excess in the transformation increased.

Although all the evidence of the present investigation was obtained from commercial carbon steels, it has been assumed, in the preceding discussion, that the conclusions arrived at may be applied, at least qualitatively, to pure iron-carbon alloys. Whether this is true, and to what extent, remains to be ascertained by future investigations, which the author cannot hope to perform. It is likely, that elements usually present in cementite or ferrite, such as manganese and silicon, also diffuse slowly in austenite at the temperatures under consideration, thus retarding the transformation. If a slow rate of diffusion in austenite of higher carbon content is the predominant factor in retarding the Ac1 transformation, it might be suggested that, in the presence of appreciable amounts of elements associated mainly with cementite, the width of the Ac1 range (reg. and cem.) would be considerably modified

[9]The effect of size and shape of the cementite particles on the velocity of the transformation has been demonstrated by Pomp and Wijkander, *Mitteilungen* aus dem Kaiser-Wilhelm Institut für Eisenforschung, 1926, Lfg. 2, Abh. 61.

[10]Loc. cit.

by variations in the fineness of cementite distribution resulting from previous heat treatment. Here the clue may be found to the different effects of manganese and silicon on the Ac1 range of carbon steels, as found by Kjerrman.[11]

2. *Spheroidizing Annealing of Carbon Steel*

Several investigators have emphasized the nucleus effect of undissolved cementite particles during the Ar1 transformation. Thus, the presence of such particles facilitates spheroidizing during annealing. From electric resistance measurements made on hypo-eutectoid steels, during heating through Ac1, coupled with annealing experiments, Kjerrman[12] concluded, that the proper annealing temperature for bringing such steels into the spheroidized condition should be "within the Ac1 interval". Judging from the results obtained by Whiteley[13] and those of the present investigation that expression would appear to require a clearer definition, in terms of microstructure.

The effects of annealing eutectoid carbon steel at temperatures:

a. coinciding with the completion of $Ac1_{reg.}$
b. above this, in the range of undissolved cementite, and
c. above this, in the range of pearlitic heterogeneity of austenite, on the resulting structure would appear to deserve being investigated.

3. *The Grain of Hardened Steel*

It is known that, for a given steel, there is a temperature range, sometimes called "hardening range", on quenching from which the most fine-grained fractures, characteristic of the steel in question, may be obtained. On exceeding that range, the fracture will be markedly coarser, the more so with increasing temperatures.

It is suggested that the presence of free carbide particles serves to obstruct grain growth in the austenite and thus indirectly is the chief agent in producing a fine-grained hardening fracture. The hardening range would then coincide approximately with the range of undissolved carbide and, in a carbon steel, depend upon

[11]Loc. cit.

[12]Jernkontorets Annaler, 1921, p. 317.

[13]Loc. cit.

carbon content and duration of heating. Possibly, the pearlitic heterogeneity will also assist in hindering the grain growth. In any case, the range of fine-grained hardening fracture would be an outcome of the slow rate of diffusion in the austenite.

The martensite, formed from austenite of pearlitic heterogeneity, is, of course, heterogeneous in composition, too. This should be one reason, why, in X-ray spectrograms obtained from such martensite, indistinct lines are always found. It may also be, that the size of the needles will be affected by the heterogeneity.

That the grain of hardened steel is refined by the presence of alloy elements, even in small amounts, is well-known; also that the hardening range is extended thereby to higher temperatures. This may be due to:

a. the finer distribution of cementite in the pearlite and thus in the austenite
b. slow rate of diffusion of the dissolved alloy elements resulting in retarded solution of the cementite.

A tentative conclusion was reached by Arpi,[14] to the effect that carbon steels that had been well deoxidized in the furnace, showed a wider hardening range than those that had not. It is conceivable that, in the latter case, the deoxidizing additions made shortly before tapping were retained in the steel largely as oxides and thus could not affect the fracture as dissolved elements would do.

4. Anomaly in Dilatation Curves Above Ac1

Andrew, Rippon, Miller and Wragg,[15] in their study of the dilatation of carbon steels of various carbon contents, found, on heating, after the contraction of Ac1, a range of rapid expansion in all steels of a carbon content exceeding about 0.6 per cent. The curve connecting the upper limits of that expansion range for different carbon contents is represented as a straight line running from about 1454 degrees Fahr. (790 degrees Cent.) for 0.80 per cent carbon to about 1607 degrees Fahr. (875 degrees Cent.) for 1.15 per cent carbon. Apparently it joins the Acm line at the latter point.

It is suggested, contrary to the opinion of those authors, that delayed solution of excess cementite in the austenite is mainly

[14] Swedish Society of Metallographers, 1927, unpublished investigation.

[15] "Effect of Initial Temperature Upon the Physical Properties of Steel." *Journal*, Iron and Steel Institute, 1920, I, p. 527.

responsible for the range of rapid expansion. It might be thought that heterogeneous distribution of dissolved carbon would also affect the dilatation curve. At any rate, that effect should be slight in comparison with the former, which accompanies a change of phase.

5. *Influence of Initial Temperature on Transformation During Cooling of Alloy Steels*

It is a well-known fact that, in certain alloy steels, as the initial temperature, from which cooling commences, is raised, lowering or "splitting" of Ar1 may be observed. Under certain conditions three definite critical points may appear. Speaking about this phenomenon, Osmond[16] stated that "there is a temperature range above Ac 3-2-1, in which the solution of the carbon on heating is incomplete, so that the carbon content is unevenly distributed in different parts of the mass. If the steel is cooled down from such a temperature, each fraction behaves as if it were alone and its recalescence occurs at a temperature which is a function of its carbon content. The frequent doubling of the recalescence point is thus explained."

The first part of this statement seems to be concurred in by several investigators and receives additional support from the present investigation. It is true, that only carbon steels have been here investigated, but it seems reasonable to expect, as suggested already, that in the presence of alloy elements the diffusion —of those elements and possibly of carbon as well—will be still more sluggish and, in consequence, the solution of the carbide present further retarded, whether this be cementite—in which alloy elements present are usually dissolved to some extent—another single carbide or double carbide. It is not necessary, for such an effect to occur, that the steel is hypereutectoid. The results reported above show clearly that the carbide of an eutectoid or even a hypoeutectoid steel may be similarly affected. For instance, a tungsten magnet steel of the usual composition, about 0.65 per cent carbon and 6 per cent tungsten, as quenched from 1560 degrees Fahr. (850 degrees Cent.), exhibits an abundance of very fine carbide particles in the structure, visible at high magnification, say 1200 diameters, as points only. Incidentally,

[16] *Revue de Métallurgie*, 1904, p. 348.

the presence of extremely fine carbide particles in the hardened structure is probably an indispensable condition for obtaining the desirable properties of a permanent magnet steel.

By raising the temperature more and more above Ac1, increasing amounts of the carbide present will be dissolved and the composition of the austenite thus modified. If the temperature is raised to a point where not only all the carbide is dissolved but also the solution has become homogeneous,[17] the matrix will have become richer in carbon and the alloy element as well and will have a composition quite different from the series of compositions present at lower temperatures. Accordingly, the transformation properties on cooling will be altered in the direction of increased sluggishness. In other words, the true transformation properties of such steels are only obtained after homogenization of the austenite at high temperature. It is not proposed to discuss this matter in detail here, since the author has done so elsewhere.[18] It may be suggested, however, that the latter part of Osmond's statement, quoted above, requires qualification in the sense that when splitting of Ar1 occurs, the upper transformation proceeds in regions appreciably larger than the small units of heterogeneity resulting from the distribution of carbide particles.

As to the nature of the three critical points on cooling the following is suggested:

Ar′ is identical with Ar1

Ar″ is initiated by separation of ferrite, so-called "secondary ferrite", in those portions of the austenite which are unaffected by Ar′

Ar‴ is the transformation into martensite of any parts left untransformed after Ar′ and Ar″. Distinct hardening is only produced by Ar‴.

It would, of course, be desirable to demonstrate the "pearlitic" heterogeneity in specimens of alloy steel quenched from moderately high temperatures above Ac1, by the use of the concentrated sodium picrate or other suitable reagent. The author's efforts to do so have, however, been so far unsuccessful, owing to the fact that the presence of this kind of heterogeneity in martensite can be

[17] Those two temperatures may or may not differ appreciably.

[18] "A Metallographic Study on Tungsten Steels," John Wiley and Sons, New York, Chapman and Hall, London, 1920, pp. 35-38.

recognized with certainty, only if the pre-existing pearlite had the typical lamellar pattern, a structure difficult to produce in high alloy steels. Furthermore, the martensitic needle-structure is apt to mask any pearlitic heterogeneity in those steels.

Among the phenomena resulting from slow diffusion of carbon in the temperature range above Ac1 the following may also be suggested:

6. the slow progress of carburization in this particular range.
7. the beneficial effect of raising the hardening temperature of such carbon and other steels as are used for dies, etc., in the highly tempered condition. In this case the solution of any hypereutectoid carbide present will add further to the effect.

Summary and Conclusions

1. The mechanism of the Ac1 transformation in carbon steels has been studied by microscopical examination of specimens hardened from within the range. It has been concluded that there are three independent factors which cause this transformation to take place in a temperature range:

a. presence of alloy elements
b. heterogeneity due to segregation on solidification
c. slow rate of diffusion of carbon and probably other elements present in austenite in a temperature range including the Ac1 range and extending beyond the latter.

2. The Ac1 range due to (1a) ($Ac1_{all.}$), in carbon steels, is narrow.

3. The Ac1 range due to (1b) ($Ac1_{seg.}$), may be wide in carbon steel that has undergone only moderate reduction in hot working [up to 85 degrees Fahr. (47 degrees Cent.) has been recorded], but only a few degrees Cent. in steels forged to small section.

4. This fact, by itself, calls for higher hardening temperatures for large objects.

5. The degree of heterogeneity due to segregation in eutectoid and hypereutectoid carbon steels may be estimated by determining the width of $Ac1_{seg.}$, due consideration being given to the condition of annealing.

6. The presence of a heterogeneous pattern, conforming with the pre-existing pearlitic structure, has been demonstrated in specimens of a 0.80 per cent carbon steel, hardened from temperatures between the beginning of the Ac1 transformation and 1472 degrees Fahr. (800 degrees Cent.), by etching in boiling concentrated sodium picrate solution. Thus, a range of so-called "pearlitic" heterogeneity in austenite has been demonstrated and point (1c) has been proved.

7. Among the phenomena that are believed to be affected by (1c) the following have been discussed or mentioned:

a. The Ac1 transformation is retarded, the more so, the coarser the cementite and the greater the amount of alloy elements present.

b. On completion of the regional Ac1 transformation, i. e., when all the ferrite of the pearlite is transformed, the average carbon content of the austenite is less than the eutectoid composition.

c. In consequence, in eutectoid carbon steel there is at this stage a considerable amount of undissolved cementite present. In the 0.80 per cent carbon steel investigated the solution of the cementite was not completed until after one hour's heating at 1418 degrees Fahr. (770 degrees Cent.).

d. The nucleus-effect of undissolved cementite, in spheroidizing annealing.

e. The temperature range, on hardening from which a fine martensitic structure and fine-grained fracture are obtained, the so-called hardening range, is believed to be related to the range of undissolved cementite. It is suggested that the grain growth in austenite and, in consequence, the development, on rapid quenching, of large martensitic needles, are hindered, when cementite particles are present.

f. The range of rapid expansion, above Ac1, noted by Andrew and co-workers, on heating eutectoid and hypereutectoid carbon steels, is believed to coincide mainly with the range of undissolved cementite.

g. It is believed, in agreement with some other investigators, that the so-called lowering or splitting of Ar1, observed

on cooling certain alloy steels from sufficiently high temperatures, is a consequence of the austenite dissolving more and more carbide and becoming homogenized, as the temperature is raised. The condition of the austenite which, on cooling, gives rise to maximum lowering, would be the homogenized one, i. e., the true condition of the steel. Lower temperatures of heating would be characterized by heterogeneity and lower alloy concentration in the austenite.

h. The slow progress of carburization at lower temperatures above Ac1.

i. The beneficial effect of high hardening temperatures when the steel is used in a highly tempered condition.

The author is indebted to the management of Söderfors Steel Works for permission to publish this paper. He gratefully acknowledges the valuable assistance of G. Lilljekvist throughout the investigation. The chemical analyses were made by E. Åberg.

Appendix

In the heating experiments described in the first part of the paper, 10 by 6 millimeter-specimens were used, the length dimension representing the radius and the flat face an axial section of the bar.

The specimens were heated in a Heræus platinum-wound horizontal tube furnace, Fig. 17. For equalization of the temperature a copper cylinder was placed in the middle of the furnace having the following dimensions: outer diameter 59 millimeters, length 90 millimeters, wall thickness 7 millimeters. Two calibrated platinum, platinum-rhodium thermocouples were employed, one inserted into the center of the copper cylinder, the other into a hole in the cylinder wall. The temperature of the latter indicated any tendency towards fluctuation in temperature, which was met by regulating the heating current.

As soon as the temperature of the test was reached, a set of specimens, bundled together in the manner shown in Fig. 17, was inserted into the cylinder, and the cylinder cover applied. It took 30 to 40 minutes to bring the temperature up to the same height again. It was then maintained for the prescribed length of time

with a variation of + 1 to — 2 degrees Cent. in the beginning, ± 0.5 degree Cent. in the latter part of the period.[19] When this was completed, the cylinder cover was removed and the bundle of specimens quickly withdrawn and quenched in water.

In order to completely remove the decarburized surface layer, 0.25 millimeter was ground off the flat surface of each specimen, which was then prepared as usual for microscopical examination. The quantity of austenite formed, in center and surface zone re-

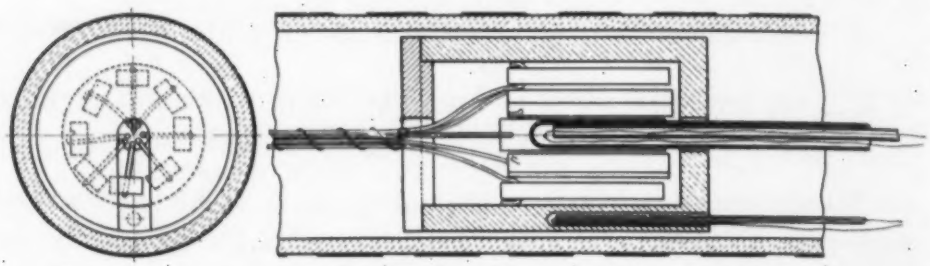


Fig. 17—Diagram of Heating Arrangement Used.

spectively, was estimated at low magnification, followed by checking of the constituents observed at high magnification. In a few cases, where some troostite had formed in quenching, special care was needed in order to distinguish this constituent from pearlite.

[19]These figures cover the variation along the length of the specimens also.

A MELTING RECORD OF THREE ACID OPEN-HEARTH HEATS

By W. E. Griffiths and C. E. Meissner

Abstract

The paper presents the complete logs of three acid open-hearth heats as made in a 25-ton acid open-hearth furnace. The records show the effect of a change in melting practice, that of adding silicon in the ladle rather than in the furnace. The effect of the various alloy additions on the slag composition and metal bath analysis is depicted graphically. A discussion of the trends of the various curves is included.

DURING 1927 the Chrome Steel Works carried out considerable experimental work to improve the analyses and quality of high-carbon acid open-hearth steel which they manufacture. The larger part of this work was carried out in the 1-ton acid open-hearth furnace which has been described by C. E. Meissner[1]. A number of experiments were carried out in the regular 25-ton acid open-hearth furnace. In the course of these investigations three preliminary heats were made to evaluate the effects of zirconium. This program entailed a decided change in melting practice, that of adding the silicon in the ladle rather than in the furnace, as is the customary procedure. With the knowledge that this change in practice would no doubt cause a considerable variation in the losses accompanying alloy additions, it was necessary that a very accurate record of the working of the heats be obtained. The data obtained from these three heats are the basis of the present paper.

These heats were made on September 13, 14 and 15, 1927, in the furnace shown in the sketch, Fig. 1. This furnace was originally designed to use producer gas for fuel but has since been changed so as to operate on oil. The method of charging was to

[1] "A One-Ton Acid Open-hearth and Some Experimental Results." Technical Publication No. 106, Class C—Iron and Steel No. 14, American Institute of Mining and Metallurgical Engineers.

A paper presented before the tenth annual convention of the society held in Philadelphia October 8 to 12, 1928. Of the authors W. E. Griffiths is research metallurgist with the Union Carbide and Carbon Research Laboratories, Long Island City and C. E. Meissner is development engineer with the Chrome Steel Works, Carteret, N. J. Manuscript received July 2, 1928.

charge first the light scrap, followed by the heavy melting scrap, car wheels and cold pig iron. The practice employed in the operation of the furnace was the standard acid open-hearth steel practice.

At intervals during the furnacing of the three heats metal and slag samples were taken from the bath of the furnace together with a record of the additions to the heat and the temperature of the slag. The metal and slag samples were carefully analyzed and the results obtained therefrom are recorded on the data tables (Tables I, II and III) together with an accurate log of the heats from the time of charging until the heats had finished pouring.

Graphs have been drawn for each heat from the chemical analyses and logs of the heats (Figs. 2 to 10). One set of curves depicts graphically the relation between the composition of the metal bath and the slag at any period during the furnacing of the heat (Figs. 2, 5 and 8). The effect of the various additions made is readily noted. There are two additional graphs for each heat; one showing the carbon, manganese and silicon content in the bath during the furnacing of the heat (Figs. 3, 6 and 9) and the other showing the elimination of iron from the slag during the working of the heat (Figs. 4, 7 and 10). A general discussion of the trend of the various elements follows.

Carbon

The carbon contents, as shown by the curves plotted from the chemical analyses, are as would be expected. The curves depict the relation of the carbon to the melting, working and finishing periods of the heats.

Heat A melted lower in carbon than did either of the other two heats. Several additions of spiegel were required to assist in cleaning up this heat previous to going ahead.

Heat B melted correctly so far as carbon content was concerned, there being sufficient carbon to allow a proper working of the heat, and as a consequence no spiegel additions were necessary.

Heat C did not melt sufficiently high in carbon and required a single spiegel addition to clean up previous to going ahead.

Manganese

Examination of the data of the log of heat A reveals a very

Table I—Log of Acid Open-Hearth Heat A

Table I—Log of Acid Open-Hearth Heat A

Time	Data	Charge Weight	Additions Weight	Metal Sample No.	Metal Sample Analysis C	Mn	Si	S	P	Cr	Zr	Slag Sample No.	Slag Sample Analysis SiO_2	FeO	MnO	Cr_2O_3	CaO	Al_2O_3	S	Fe_2O_3	MgO	Fe_2O_3	Total Fe	Total	Metal in slag Si	Mn	Direction of Gas R	L	Temperature Optical Pyrometer Observed	True
8:05	Begin Charging	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
-	Pig Iron	10800	-	7244	(CC 3.28 (CC .61 (TC 3.89	.63	2.13	.023	.032	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
-	Car Wheels	25100	-	-	{ .95 { .70	.72 .66	- .28	.045 .048	.009 .045	- .07	- -	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
-	Chrome Scrap	9600	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
-	Heavy Melting	4100	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
-	Total Charge	50600	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
11:00	Finish Charging	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
11:30	1st Pool Metal	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
11:34	Direction Gas	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	→	-	-	-
11:54	Peels front all doors	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	→	-	-	-
12:15	Slag Sample A	-	-	-	-	-	-	-	-	-	-	7251	42.18	46.10	11.21	1.62	.10	1.00	.124	-	Nil	2.83	33.00	99.05	-	-	→	-	-	-
12:30	Slag Sample B	-	-	-	-	-	-	-	-	-	-	7252	42.50	40.50	11.74	1.74	.10	1.00	.060	-	Nil	1.92	32.74	99.45	-	-	→	-	-	-
12:45	Slag Sample C	-	-	-	-	-	-	-	-	-	-	7253	43.18	40.00	11.79	1.64	.01	1.00	.092	-	Nil	1.57	32.26	99.51	-	-	→	-	-	-
12:56	Metal nicely under) cover,Boil vigorous)	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
12:55	Metal & Slag Samples 1	-	-	7262	.42	.041	.06	.042	.038	.023	-	7254	44.20	39.80	11.90	1.61	.01	1.00	.076	-	Nil	.84	31.51	99.56	-	-	-	-	-	-
1:30	Metal & Slag Sample 2	-	-	7263	.68	.07	.10	.036	.034	.025	-	7256	49.40	33.60	12.31	1.30	.01	1.00	.120	-	Nil	1.29	27.0	99.20	-	-	-	-	-	-
1:35	Spiegel #3 Door	-	100	-	3.99	19.86	1.38	.006	.076	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
1:49	Metal & Slag Sample 3	-	-	7264	.69	.10	.09	.031	.039	.06	-	7256	48.20	34.20	12.16	1.24	.01	1.00	.092	-	Nil	8.13	28.0	99.18	-	-	-	-	-	-
	Limestone #3 Door																							(CO2 + Organic)						
1:55	Violent	-	400	-	-	-	-	-	-	-	-	7245	1.23	-	.05	-	53.54	.40	.034	.014	.99	.40	45.08	99.54	-	-	-	-	-	-
2:05	Temperature	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	off	-	2949°F	-
2:08	Metal & Slag Sample 4	-	-	7265	.69	.06	.05	.041	.038	.08	-	7257	51.12	28.20	12.51	1.10	3.48	1.00	.104	-	Nil	1.47	22.94	99.14	-	-	-	-	-	-
2:25	Temperature	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	off	-	2988°F	-
2:25	Spiegel	-	300	-	3.99	19.86	1.38	.006	.076	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
2:43	Metal & Slag Sample 5	-	-	7266	.67	.10	.04	.040	.036	.051	-	7258	54.19	21.70	14.13	.83	3.16	1.00	.18	-	Trace	4.19	19.79	99.65	.05	.08	-	-	-	-
	Very fierce M.C. Ferrochrome																													
2:45	#3 Door	-	560	7268	5.06	.48	-	.11	.007	58.68	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
2:50	Temperature	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	off	-	3001°F	-
2:56	Metal & Slag Sample 6	-	-	7267	.56	.10	.05	.033	.040	.63	-	7259	56.36	18.60	14.22	2.97	2.67	1.50	.10	-	Trace	3.75	17.06	99.29	-	-	-	-	-	-
3:00	Gas off	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	off	-	-	-
3:05	Ferromanganese	-	400	7247	7.15	79.92	.25	.03	.121	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
3:10	Temperature	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	off	-	2829°F	-
3:15	Temperature	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	off	-	2790°F	-
3:18	Metal & Slag Sample 7	-	-	7268	.61	.37	.07	.038	.042	.67	-	7260	56.00	13.30	17.10	4.31	2.98	1.70	.092	-	Trace	3.56	12.82	99.18	-	-	-	-	-	-
3:35	Metal & Slag Sample 8	-	-	7269	.58	.26	.07	.035	.039	.67	-	7261	56.54	12.30	18.90	4.31	3.10	1.00	.076	-	Trace	2.95	11.61	99.29	.09	.09	-	-	-	-
3:35	Heat - Cold	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
3:36	Gas on	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	On	-	-	-
3:38	Tapped Heat	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
	Ladle Additions	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
	Silicon-Zirconium	-	365	-	-	-	43.29	-	-	-	11.78	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
	Aluminum	-	5	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
	Temperature Tapping	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	2661°F	2895°F
3:42	Ladle Pulled	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
3:44	Begin Pouring	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	2547°F	2770°F
	Temperature 12th Mold	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	2529°F	2740°F
3:50	Ladle Sample	-	-	7271	.59	.24	.30	V .038 G .049	.042	.73	.047	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
4:34	Finished Pouring	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	2525°F	2760°F
	Total Metallic Charge	52328	1728																											

Ingots	21541
Foundry	21188
Estimate Skull	1500
Scrap	6577
Estimate Furnace Loss	1525
Oil Used	960 Gal.

DATA

Small Tap Hole
Ladle 1-1/2" Nozzle
Slag kept in furnace
Furnace bottom good
V = Volumetric Analysis
G = Gravimetric Analysis

ADDITION CALCULATIONS

	C	Mn	Si	P	S	Cr	Zr
Sample 5	.67	.10	.04	.036	.040	.03	
Ferrochrome	.06	.006	-	-	-	.74	-
Ferromanganese	.05	.61	-	-	-	-	-
Silicon-Zirconium	-	-	.31	-	-	-	.083
Total	.47	.71	.56	-	-	.77	.083
Final Analysis	.59	.24	.30	.042	.043	.73	.047
Losses	.08	.47	.06	-	-	.04	.036

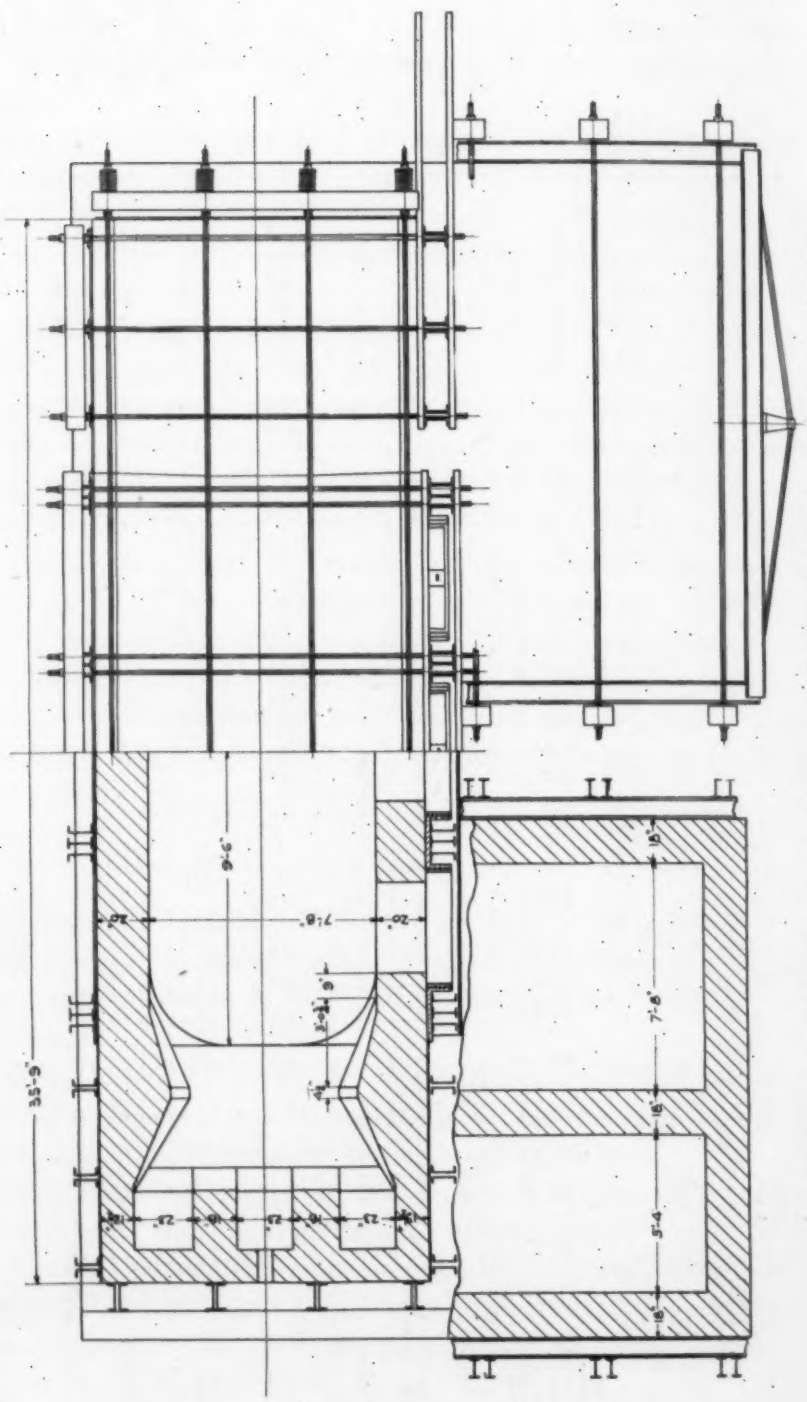


Fig. 1—Plan and Section of Open-Hearth Furnace Used in Making the Heats Described in this Paper.

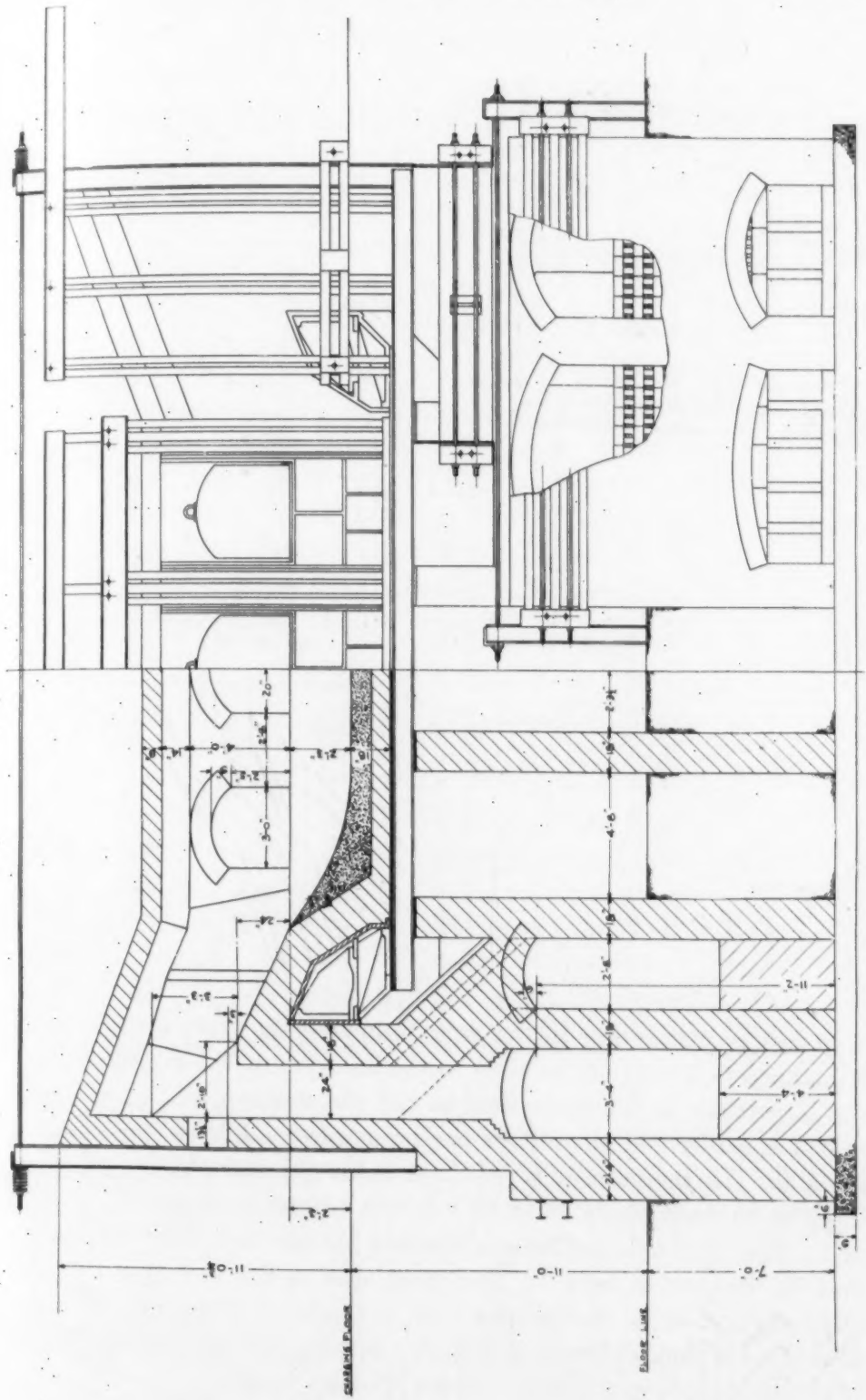


Fig. 1a—Elevation and Section of Acid Open-Hearth Furnace Used in Making the Heats Described in this Paper.

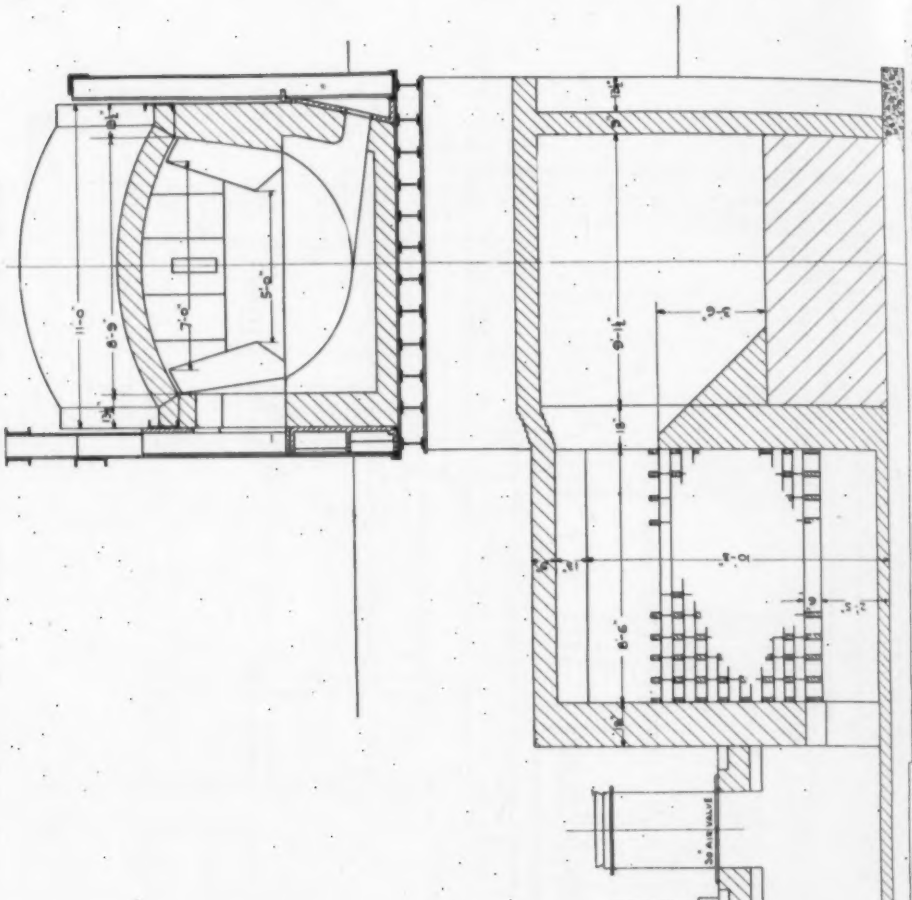


Fig. 1b—End View and Section of Acid Open-Hearth Furnace Used in Making the Heats Described in this Paper.

high loss in manganese in the bath following the addition of ferro-manganese. Attention is called to the analyses of the samples from the slag taken at the same time as the metal samples. They show that a considerable amount of the manganese addition was lost to the slag. In the case of these three heats the manganese was put into the furnace about 16 to 33 minutes before the heats were tapped. This may account to some extent for the high loss of manganese in the case of heat A. However, due to the fact that heat B had only a normal manganese loss, a point more worthy of consideration is the difference in practice employed in the manufacture of the zirconium and non-zirconium treated heats.

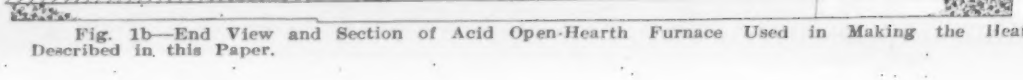


Fig. 1b—End View and Section of Acid Open-Hearth Furnace Used in Making the Heats Described in this Paper.

Table III—Log of Acid Open-Hearth Heat C

Time	Data	Charge Weight	Additions Weight	Metal Sample No.	Metal Sample Analysis C	Mn	Si	S	P	Cr	Slag Sample No.	Slag Sample Analysis SiO_2	FeO	MnO	Cr_2O_3	CaO	Al_2O_3	S	P_2O_5	MgO	Fe_2O_3	Total Fe	Total	Metal in Slag Si	Mn	Direction of Gas R	L	Temperature Optical Pyrometer Observed	True
8:25	Began Charging	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
					DC 3.28																								
-	Pig Iron	11800	-	7244	DC .61	.63	2.13	.023	.031	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
					DC 3.89																								
					.93	.72	-	.046	.009	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
-	Car Wheels	22000	-	-	.70	.65	.28	.048	.065	.07	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
-	Chrome Scrap	11900	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
-	Heavy Melting	5100	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
-	Total	50800	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
9:30	Finished Charging	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	←	-	-	-
11:36	Sample A - Not all melted.	-	-	7272	.11	.04	.02	.052	.061	.07	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	→	-	-	-
12:00	Charge Gas	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
12:15	Sample 1	-	-	7273	.36	.04	.08	.027	.076	.08	7284	44.30	41.00	8.83	1.73	Trace	1.22	.120	-	None	1.95	33.28	99.33	-	-	-	-	-	-
12:45	Sample 2	-	-	7274	.53	.07	.09	.050	.070	.16	7285	44.74	39.00	9.87	2.17	Trace	1.40	.136	-	None	1.69	31.48	99.21	-	-	-	-	-	-
1:15	Sample 3	-	-	7275	.86	.12	.18	.042	.073	.11	7286	47.24	35.40	10.39	2.19	.20	1.30	.144	-	None	2.35	29.14	99.45	-	-	-	-	-	-
2:00	Heat melted Good slag action	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	off	-	3027°F	-
2:00	Sample 4	-	-	7276	1.02	.14	.26	.042	.068	.11	7287	48.44	34.70	10.36	2.12	.20	1.00	.124	-	Trace	1.93	28.31	99.25	-	-	←	-	-	-
2:30	Sample 5	-	-	7277	1.00	.13	.23	.041	.073	.12	7288	49.20	33.58	10.77	2.04	.84	1.00	.092	-	Trace	1.58	27.20	99.26	.09	.10	-	-	-	-
2:30	Limestone-Temperature	-	400	-	-	-	-	-	-	-	7243	1.23	-	.03	-	53.34	.40	.024	.014	.99	.40	*43.08 + CO_2 + Organic	99.64	-	-	off	2936°F	-	-
2:56	Sample 6	-	-	7278	.83	.07	.06	.035	.068	.16	7289	56.42	13.35	11.75	1.85	6.10	1.56	.076	-	Trace	8.44	16.25	99.52	-	-	-	-	-	-
3:00	Temperature	-	-	-	-	-	-	-	-	-	7243	1.23	-	.03	-	53.34	.40	.024	.014	.99	.40	*43.08 + CO_2 + Organic	99.64	-	-	off	3039°F	-	-
3:06	Limestone	-	200	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
3:16	Sample 7	-	-	7279	.73	.06	.05	.043	.066	.16	7290	57.36	16.49	10.96	1.39	6.80	1.70	.124	-	Trace	5.93	16.19	99.67	.07	.09	-	-	-	-
3:30	Temperature	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	off	3046°F	-	-
3:45	Sample 8	-	-	7280	.58	.07	.06	.044	.067	.17	7291	58.86	15.56	8.08	1.30	7.48	1.80	.136	.016	-	5.59	16.00	99.02	-	-	off	2930°F	-	-
3:46	Ferrosilicon #1 Door	-	300	7345	-	-	47.96	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
3:50	H.C. Ferrochrome #1 Door	-	500	7349	6.02	.48	-	.11	.006	68.66	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
3:55	H.C. Ferrochrome #1 Door	-	60	7348	6.02	.48	-	.11	.006	68.66	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
	Slag quieting down -	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
	still some action -	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
	Slag is thicker	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
4:05	Gas Off	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	off	3040°F	-	-
	Temperature	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	off	2923°F	-	-
4:05	Ferromanganese #2 Door	-	400	7347	7.20	79.92	.25	.003	.120	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	off	2868°F	-	-
4:07	Temperature	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	off	2780°F	-	-
4:20	Temperature	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
4:28	Sample 9	-	-	7281	.65	.89	.37	.041	.068	.93	7292	57.56	17.72	11.11	2.22	6.20	1.60	.076	.016	-	2.46	15.49	99.25	-	-	off	2742°F	-	-
4:33	Temperature	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	2698°F	2965°F	-
4:36	Heat Tapped	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
	Aluminum	-	3	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
4:46	Ladle Pulled	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	2619°F	2845°F	-
	Temperature 1st Mold	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	2591°F	2825°F	-
	Temperature 4th Mold	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	2540°F	2770°F	-
4:50	Temperature Mold	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	2510°F	2720°F	-
4:51	Temperature Ingot	-	-	7282	.64	.54	.36	.053	.074	.90																			
	1st Ladle Sample	-	-	7283	.61	.54	.36	.049	.073	.91																			
	2nd Ladle Sample	-	-																										
	Total Charge	[illegible]	1560																										

Weight Ingots	43510
Weight Foundry	3996
Weight Scrap	1647
Furnace Loss	2510
Oil Used	1050 gal.

DATA

Condition of slag at tap - Dark green - Good
Condition of metal at tap - Good
Condition of metal in molds - Good
Condition of furnace bottom - Good
Size Ladle Nozzle - 1-3/4"
Skull - None

ADDITION CALCULATIONS

Sample No.	C	Mn	Si	S	P	Cr
5-7280	.58	.07	.06	.044	.067	.17
Ferrosilicon	-	-	.28	-	-	-
Ferrochromium	.064	.006	-	-	-	.74
Ferromanganese	.066	.61	.008	-	.001	-
Total Points Added	.69	.68	.34	.044	.068	.91
Final Analysis	.61	.54	.36	.049	.073	.91
Losses	.08	.14	+.02	+.005	+.005	.00

Table II—Log of Acid Open-Hearth Heat B

Time	Data	Charge Weight	Additions Weight	Metal Sample No.	Metal Sample Analysis C	Mn	Si	S	P	Cr	Zr	Slag Sample No.	Slag Sample Analysis SiO_2	FeO	MnO	Cr_2O_3	CaO	Al_2O_3	S	P_2O_5	MgO	Fe_2O_3	Total Fe	Total	Metal in Slag Si	Mn	Direction Gas R	L	Temperature Optical Pyrometer Observed	True
0:45	Began Charging	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
-	Pig Iron	11400	-	7344	(cc .61 (cc 3.88 (cc 4.49	.63	2.13	.025	.051	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
					(.95	.72	-	.048	.009	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
					(.70	.66	.89	.048	.048	.07	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
-	Car Wheels	24300	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
-	Chrome Scrap	14500	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
-	Total Charge	52200	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
10:00	Finished Charging	-	-	7303	.16	.06	.03	.036	.026	.06	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
11:30	Metal Sample A -] first pool of metal)	-	-	7304	.25	.06	.03	.036	.030	.10	-	7295	45.80	36.91	10.61	2.63	.39	1.80	.18	-	Trace	1.14	31.05	99.83	-	-	-	-	-	-
12:00	Sample 1	-	-	7305	.42	.06	.04	.031	.032	.12	-	7294	46.36	34.87	11.10	2.88	.56	1.80	.06	-	None	.81	27.0	99.84	-	-	-	-	-	-
12:30	Sample 2	-	-	7306	.64	.08	.09	.030	.027	.13	-	7295	48.08	34.89	11.18	2.78	.61	2.00	.06	-	Trace	.83	27.0	99.23	.40	.08	←	-	-	-
1:00	Sample 3	-	-	7307	.62	.10	.16	.030	.031	.12	-	7296	47.42	34.01	11.36	2.54	.94	2.00	.086	-	Trace	1.23	26.72	99.35	-	-	→	-	-	-
1:30	Sample 4	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
1:50	Charge all melted. Lots of action in bath.	-	-	7308	.93	.11	.19	.028	.033	.12	-	7297	47.84	33.29	11.48	2.41	.96	2.00	.066	-	Trace	1.25	26.72	99.35	-	-	-	-	-	-
2:00	Sample 5	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
2:10	Gas change every 20 mins.	-	400	7245	-	-	-	-	-	-	-	-	1.85	-	.03	-	53.34	.40	.084	.014	.99	.40	-	(43.08 Organic Matter)						
2:20	Limestone @ Door	-	-	7309	.88	.08	.12	.029	.034	.11	-	7298	50.47	28.83	11.38	2.19	4.64	1.10	.06	-	Trace	.89	22.89	99.25	.10	.04	-	-	-	-
2:30	Sample 6	-	-	7310	.80	.06	.06	.030	.051	.15	-	7299	54.44	22.03	12.89	1.81	4.77	1.00	.06	-	Trace	2.97	19.21	99.46	-	-	-	-	-	-
2:55	Sample 7	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
2:56	Lots of action in bath - Sticky Slag	-	300	7349	3.99	19.86	1.38	.006	.076	.16	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	Off	-	3064°F	-
3:00	Spiegel	-	-	-	-	-	-	-	-	-	-	-	1.85	-	.03	-	53.34	.40	.084	.014	.99	.40	-	(43.08 Organic Matter)						
3:10	Temperature	-	200	7243	-	-	-	-	-	-	-	7300	57.52	19.17	15.00	1.61	6.08	.84	.066	-	Trace	1.43	15.86	99.79	-	-	-	-	-	-
3:15	Limestone	-	-	7311	.72	.09	.05	.033	.034	.16	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
3:30	Sample 8	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
3:50	Lots of action in bath- Fluid slag	-	-	7312	.61	.07	.04	.038	.031	.16	-	7301	57.34	13.60	13.12	1.43	5.84	1.80	.076	.025	-	7.08	15.49	99.85	-	-	-	-	2969°F	-
3:60	Sample 9	-	560	7248	5.02	.48	-	.11	.006	68.66	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	Off	-	-	-
3:68	Ferrochrome	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	Off	-	-	-
4:10	Temperature	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	Off	-	2780°F	-
4:10	Gas Off	-	400	7247	7.80	79.92	.25	.003	.180	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
4:12	Ferromanganese	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
4:25	Temperature	-	-	7313	.65	.41	.08	(*.029 (.034	.051	.82	-	7302	57.00	12.70	13.36	4.76	5.86	1.80	.068	.023	-	4.46	12.99	99.85	-	-	-	-	2724°F	2970°F
4:35	Sample 10	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
4:28	Tapped Heat	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
	Temperature] Metal in runner]	-	-	-	-	-	43.29	-	-	-	11.76	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	2864°F	2820°F
	Silicon-Cr. in ladle	-	365	7246	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	2862°F	2790°F
	Aluminum in ladle	-	5	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
4:38	Ladle pulled	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
	Temperature - 1st mold	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-
	Temperature - 15th mold	-	-	7314	.67	.36	.34	**.044	.040	.80	.047																			
	1st Ladle Sample	-	-	7315	.66	.37	.34	**.014	.040	.80	.047																			
	2nd Ladle Sample	-	-																											
	Total Heat) Charged	51878	1828																											

Weight of Ingots	34665
Weight to Foundry	13986
Weight Scrap	994
Furnace Loss	2183

Size Ladle Nozzle 44/8"
Oil Used - 1040 gal. containing .32% by weight of sulphur.

NOTE: * - .029 Sulphur Volumetric dilute acid
.034 " " concentrated acid
** - Gravimetric Analysis
Check - .041 Volumetric Concentrated Acid
.040 " "
.047 Gravimetric method

ADDITIONS CALCULATIONS

Sample	C	Mn	Si	P	S	Cr	Zr
7312-9	.61	.070	.040	.031	.038	.16	-
Ferrochromium	.05	.006	-	-	-	.74	-
Ferromanganese	.08	.610	.003	.001	-	-	.088
Silicon-Zirconium	-	-	.305	-	.032	.038	.90	.088
Total Points Added	.71	.68	.34	.032	.038	.90	.088
Final Analysis	.66	.37	.34	.040	.043	.80	.067
Losses	.06	.31	.00	+.008	+.005	.10	.034

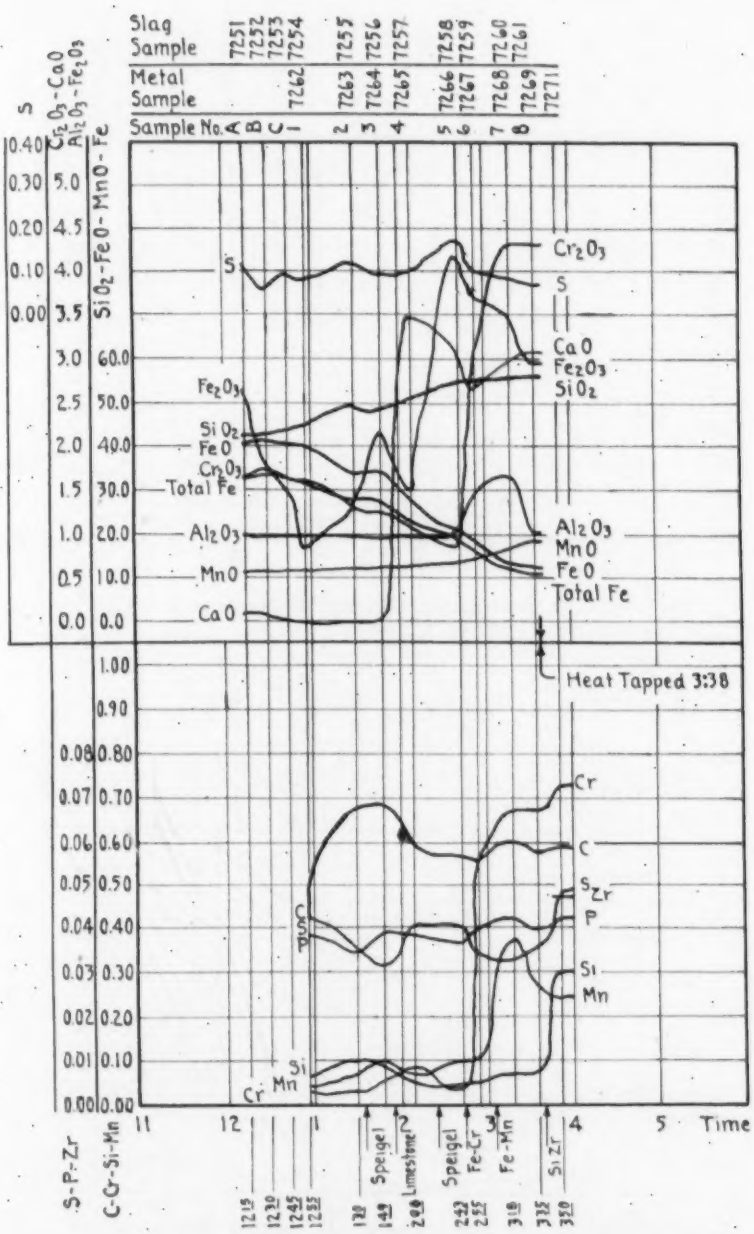


Fig. 2—Curves Showing the Relation Between the Composition of the Metal Bath and the Slag at any Period During the Furnacing of Heat A.

At the time of the manganese and chromium additions to heats A and C there had been no silicon added to the bath except that contained in the spiegel, therefore, the metal would not be in the same condition as it is in the usual practice. There is no doubt but that it contained a sufficiently greater amount of oxide to cause the

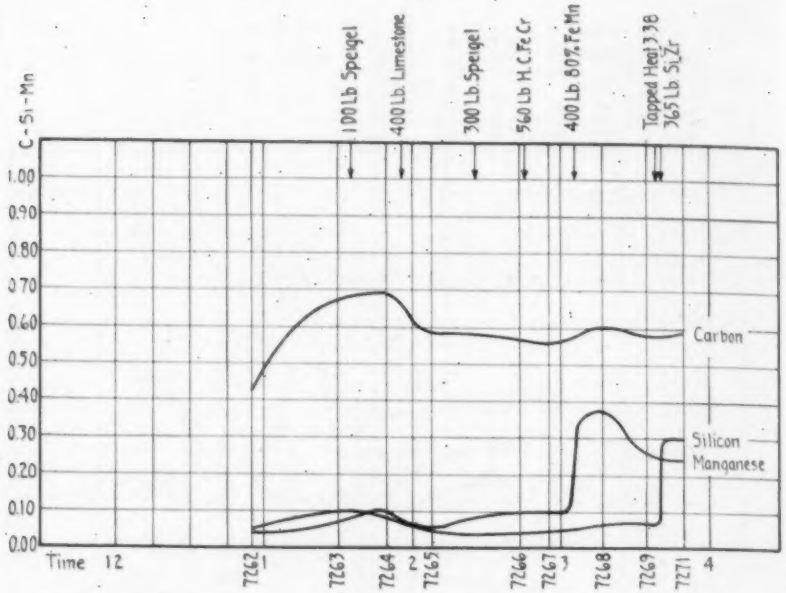


Fig. 3—Curves Showing the Carbon, Manganese and Silicon Content in the Bath During the Furnacing of Heat A.

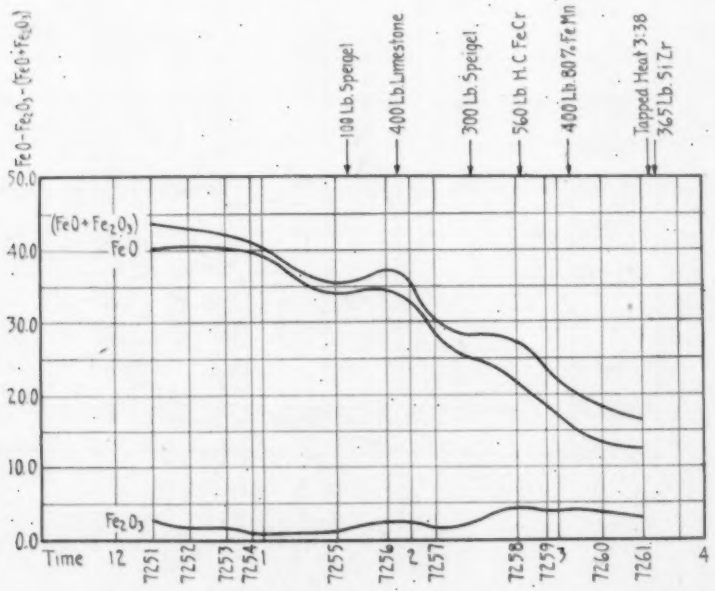


Fig. 4—Curves Showing the Elimination of Iron from the Slag During the Working of Heat A.

increased manganese loss, as was obtained. The fact that the chromium losses were not exceedingly great is not important, inasmuch as manganese is more readily oxidized than chromium when present in the amounts under consideration. The losses in the case of heat B, made by the usual practice, in which the silicon was added in

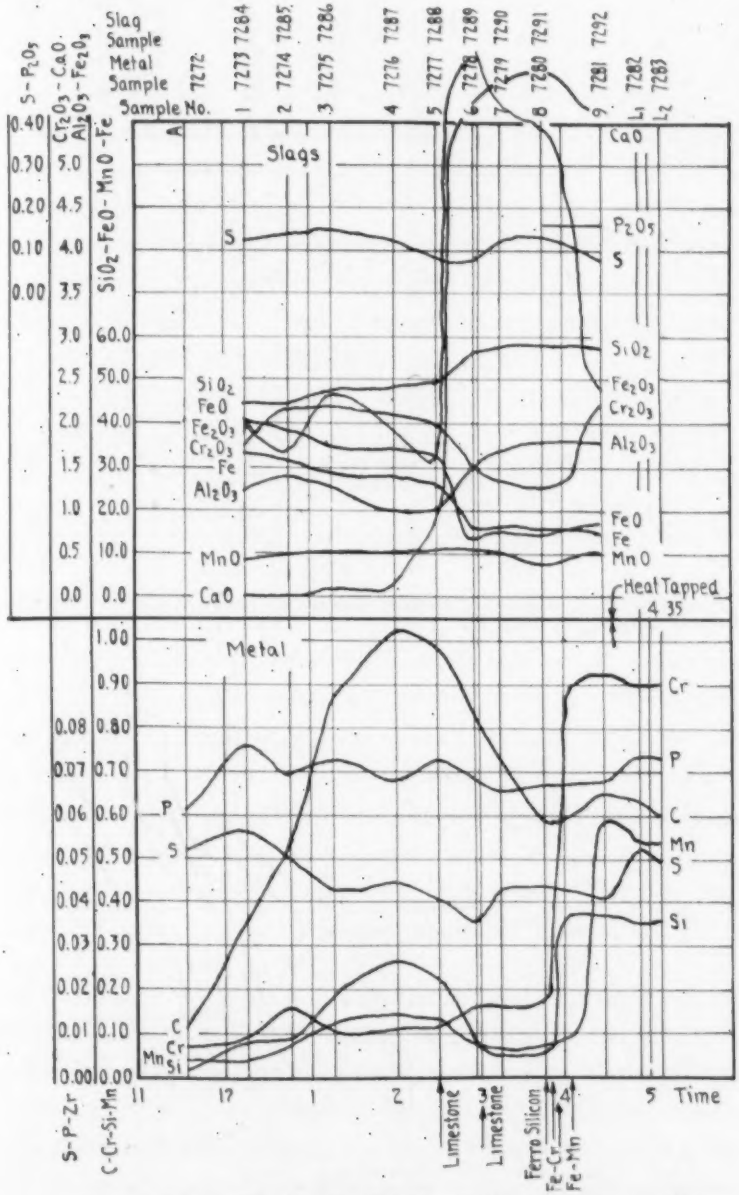


Fig. 5—Curves Showing the Relation Between the Composition of the Metal Bath and the Slag at any Period During the Furnacing of Heat B.

the furnace previous to the addition of manganese and chromium, were such as would be expected in regular practice. It was deduced, from the data tabulated, that a proper amount of ferrosilicon should be added in the furnace previous to the chromium and manganese additions if it is desired to avoid a considerable loss in these

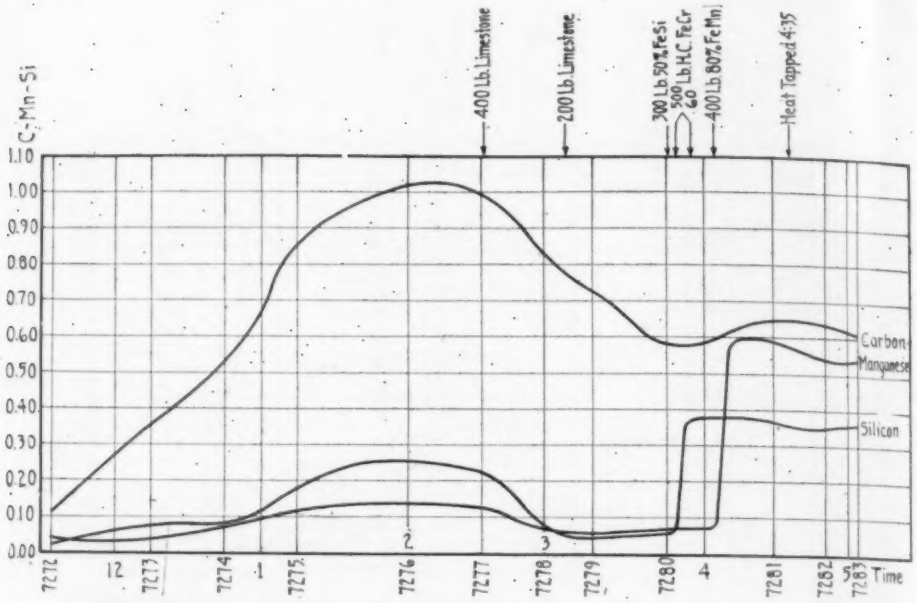


Fig. 6—Curves Showing the Carbon, Manganese and Silicon Content in the Bath During the Furnacing of Heat B.

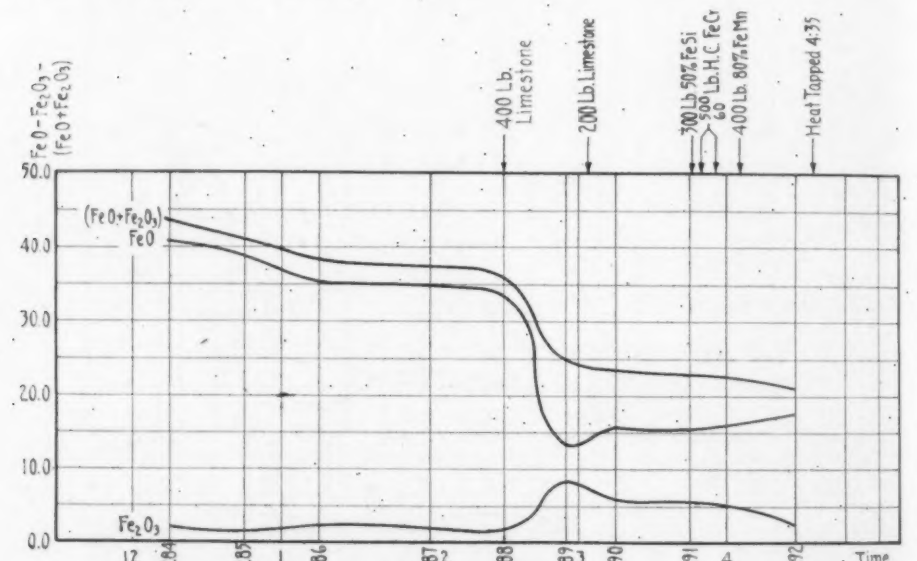


Fig. 7—Curves Showing the Elimination of Iron from the Slag During the Working of Heat B.

elements. As will be noted under the discussion of silicon, the furnace as operated did not take up silicon from the bottom. Therefore, since the silicon and manganese contents were very low at the beginning of each heat no doubt the bath contained a considerable amount of oxide, which accounts for the high manganese loss.

Silicon

There was no loss in silicon from the silicon addition in the case of heats B and C. However, heat A lost 0.05 per cent silicon. It is believed that this loss was due to a furnace condition. The furnace was very hot at the time the calculation for the silicon addition was made and in allowing the bath to cool by shutting off the gas, the metal in cooling lost some of its silicon to the slag.

As was noted in the discussion of manganese, the curves show that the furnace was operated in such a manner that it did not pick up any silicon from the bottom. This point is important because where a furnace is operated in such a manner that the bath does have a silicon pick-up, the losses from the various additions would no doubt be considerably less.

Chromium

In the case of the heats made where the silicon was added as a ladle addition there is a slightly higher loss in chromium than is obtained when the ferrosilicon is added in the furnace previous to the chromium addition. The curves show a slight increase in chromium oxide in the slag with the addition of ferrochromium. This is the expected result.

Phosphorus

The phosphorus content of heats A and C was decidedly lower throughout the entire furnacing of the heats than that of heat B. This condition was no doubt due to the higher phosphorus content in the scrap charge in the case of heat B. As will be noted, there is no elimination of phosphorus in the three heats under consideration. This is entirely in agreement with the data on acid open-hearth practice.

Sulphur

It is a well established fact that acid open-hearth steel practice does not eliminate sulphur. Examination of the sulphur determinations as plotted check this statement. The curves have been plotted to show the highest sulphur content in the ladle samples. In the case of the ladle samples from heats A and C the sulphur

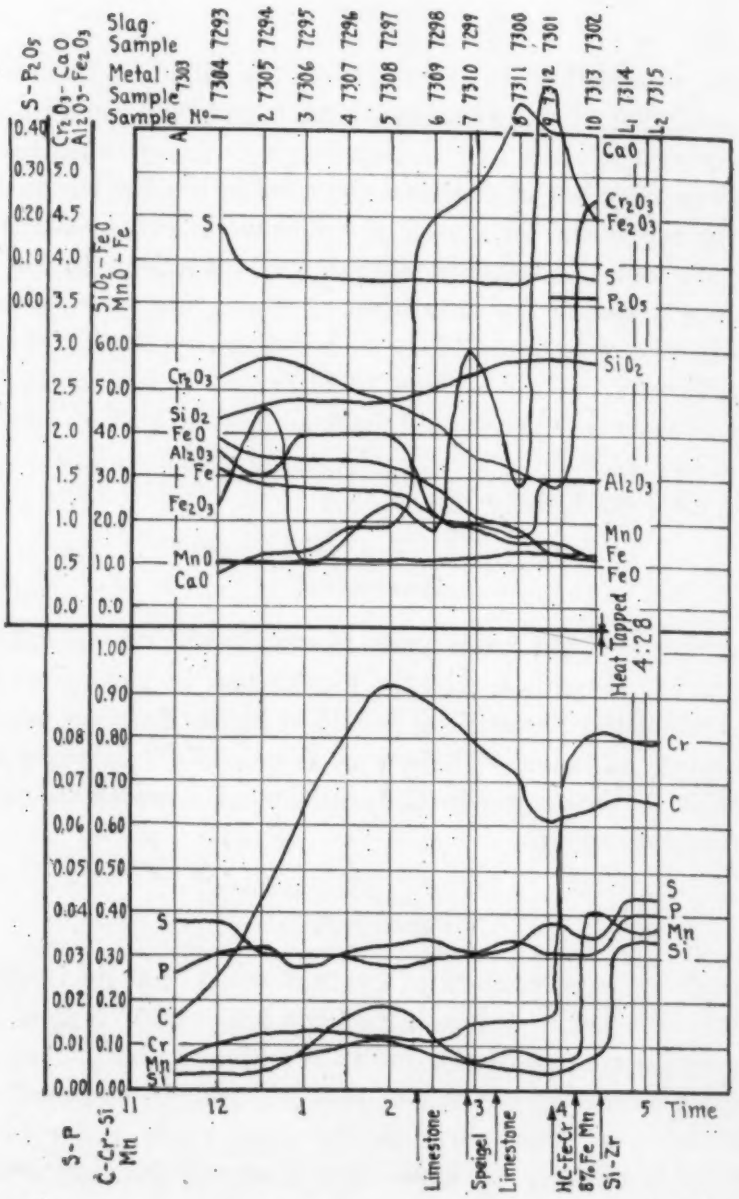


Fig. 8—Curves Showing the Relation Between the Composition of the Metal Bath and the Slag at any Period During the Furnacing of Heat C.

content was determined gravimetrically. If the volumetric analyses were plotted the sulphur content would be slightly lower. This condition is due, in the case of heats A and C, to the fact that zirconium combines chemically with the sulphur and forms an insoluble sulphide which is not acted upon by 1:1 hydrochloric acid, used in

volumetric determination. Sulphur present in this form is not detected by nor included in any analysis performed by the ordinary evoluton method. In the gravimetric method involving oxidation, however, it is quantitatively determined. A. L. Feild[2] in one of his papers establishes a relationship between the percentage of sulphur "fixed" and the zirconium added. This relationship is expressed as:

$$\text{Per cent sulphur ''fixed''} = \frac{\text{Per cent added zirconium}—0.15}{10}$$

A more recent equation[3] is:

$$\text{Per cent sulphur ''fixed''} = \frac{\text{Per cent added zirconium}—(0.15 + 1.418)}{10}$$

S = sulphur

If we substitute the percentage of added zirconium in the case of the two heats made in the Chrome Steel Works, according to either formula, a negative quantity of sulphur would be "fixed." For example:

$$\text{Per cent sulphur ''fixed''} = \frac{0.082—0.15}{10} = \text{negative quantity}$$

It would therefore be concluded that there was not sufficient zirconium added to these heats to "fix" any of the sulphur. This does not seem to be the case. The sulphur analysis for heat A shows that a small amount of sulphur was apparently "fixed" by the addition of silicon-zirconium. The volumetric chemical analysis, using concentrated acid for the ladle sample of heat A shows 0.035 per cent sulphur, whereas the gravimetric analysis shows 0.048 per cent sulphur. It is concluded, therefore, that 0.013 per cent sulphur was "fixed" by the addition of the silicon-zirconium.

In the case of heat B gravimetric sulphur was not determined. The sulphur content of this heat is slightly higher than that of either heat A or C. This condition is no doubt due to the higher sulphur content in the scrap charge.

[2]Alexander L. Feild, "Some Effects of Zirconium in Steel," *Transactions*, American Institute of Mining and Metallurgical Engineers, 1923.

[3]Alexander L. Feild, "Effect of Zirconium on Hot Rolling Properties of High Sulphur Steels and Occurrence of Zirconium Sulphide," *Transactions*, American Institute of Mining and Metallurgical Engineers, 1924.

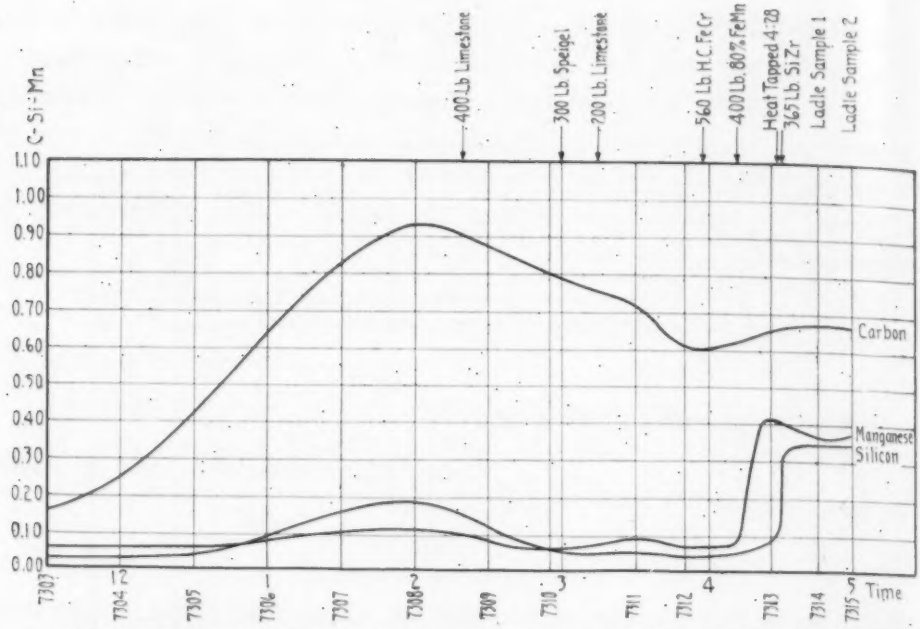


Fig. 9—Curves Showing the Carbon, Manganese and Silicon Content in the Bath During the Furnacing of Heat C.

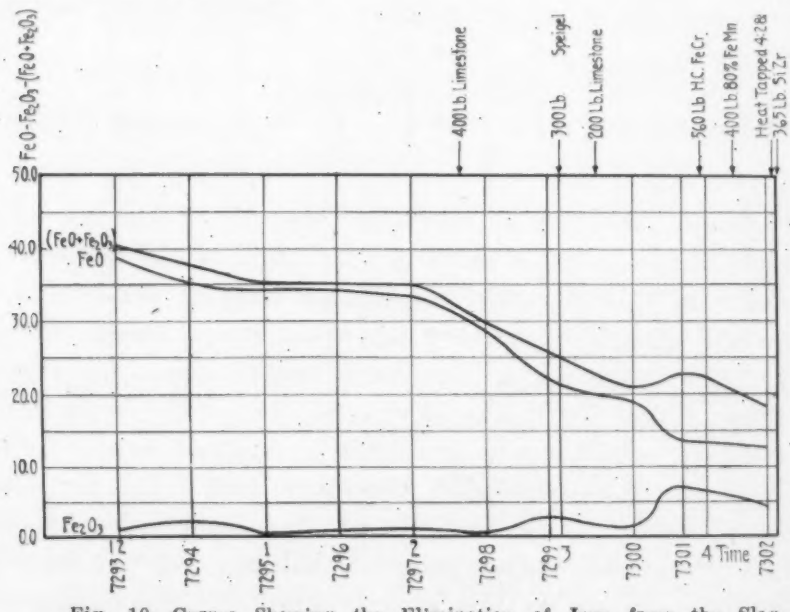


Fig. 10—Curves Showing the Elimination of Iron from the Slag During the Working of Heat C.

Heat C, Sample 7313, taken previous to the addition of silicon-zirconium, shows 0.034 per cent sulphur as determined by using concentrated acid and the evolution method. The first and second ladle samples secured after the zirconium addition show 0.044 and

0.047 per cent sulphur as determined by gravimetric analysis. The volumetric analysis of the ladle samples, using concentrated acid, shows 0.041 per cent sulphur. It appears, therefore, that at least 0.006 per cent sulphur was "fixed" by the addition of zirconium.

It is probable that the "fixing" of the sulphur by such a small addition of zirconium, is accomplished through the manufacture of the heats in an acid open-hearth furnace.

Slag Components

The analyses and curves plotted therefrom for calcium oxide and chromium oxide, show an increase in content of these constituents in a direct relation with the lime and chromium additions.

The silica (SiO_2) analyses and curves plotted therefrom show the silica increasing in almost a straight line function as the heat is worked and finished in the furnace.

The iron in the slag present as iron oxides (FeO and Fe_2O_3) decreases in a regular manner from the melting period to the finishing period. Once the heat is all melted and the elimination of the carbon has commenced the slope of the iron oxide curve is a good bit steeper than that at any other period during the furnacing of the heat. This shows that during this period the iron oxide is being reduced quite rapidly. The curves showing the relation of iron oxide present as Fe_2O_3 and FeO during the various periods throughout the heat show that when there is a decided drop in FeO there is a corresponding increase in Fe_2O_3. When the two components are reduced to a single constituent and plotted, the slope of the curve is quite regular, the amount gradually decreasing as the heat is finished.

A very interesting relation is to be noted between the iron oxides present in the slag and the chemical analyses of the bath samples with respect to losses from the ferro alloy additions.

In the case of heat A the elimination of iron from the slag continues after going ahead with the heat. In other words, the slope of the curve is such that it appears that the heat was not ready to go ahead. The heavy loss in ferromanganese and slight loss in ferrochromium are probably due to this condition.

Heat C shows a somewhat similar condition, although the heat was apparently in better condition before going ahead, as is shown

by the curve plotted for the total oxides on the graph of heat C.

Heat B was apparently in better shape than either of the other two heats before going ahead. It is to be noted that the total iron oxide curve for the slag of this heat was approximately horizontal during the finishing stages of the heat and had reached this position previous to going ahead with the heat.

Heat A as tapped from the furnace had no slag in the ladle. This condition was due to the turning off of the gas about 35 minutes before tapping which caused the slag to chill over and be retained in the furnace in tapping.

The slags contained small metal particles which were in several samples analyzed separately. The results are shown on the table following:

Heat No.	Slag Sample	Ratio of Slags to Metallic	Manganese	Silicon
"A"	5	317:7	0.08	0.05
	8	385:10	0.09	0.09
"B"	5	250:3	0.10	0.09
	7	342:19	0.09	0.07
"C"	4	320:2.15	0.05	0.40
	6	325:11	0.04	0.10

Temperature

The temperature observations were made with an optical pyrometer. This instrument had been recently standardized by the manufacturer and it is believed that the temperature readings are accurate. There is, of course, a human element (the operators) which enters into the taking of an observation of this kind, but this should be of little consequence as various operators checked the readings. The readings made in the furnace do not require a correction, the observed temperature is the true temperature. The readings were made during reversals when the flame was off. The true pouring temperatures were obtained from a curve furnished by the manufacturer.

This paper is presented to the Steel Melting Session as a contribution to the literature on acid open-hearth steel melting practice. The authors feel that it offers valuable information to those actively engaged in the manufacture of such acid steels.

The authors wish to acknowledge the courtesy of the Chrome Steel Works and the Union Carbide and Carbon Research Laboratories in permitting this data to be published.

We also wish to acknowledge the assistance of A. L. Feild of the Electro Metallurgical Co., who directed certain phases of this work, J. B. Emory and W. A. Quadenfield of the Chrome Steel Works, who collaborated in making the observations, and Superintendent Alberts, who was in charge of the furnace. To the Union Carbide and Carbon Research Laboratories chemical staff is given the credit for making the painstaking chemical analyses.

DISCUSSION

A. L. Feild: Union Carbide and Carbon Corp., New York City.

These heats of which we find a record in this paper were made in part to determine the effect of making silicon additions in the ladle on recoveries. The necessity for making the silicon addition in the ladle arose from the fact that there was used in two of these heats a silicon-zirconium alloy whose zirconium content is lost very rapidly when the addition is made in the furnace.

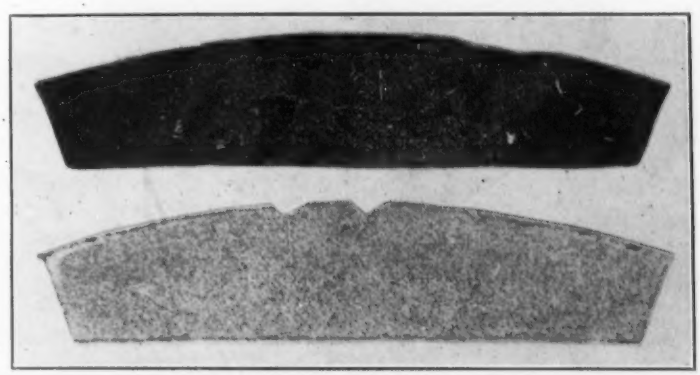


Fig. 1—Sulphur Prints Made Under Strictly Comparable Conditions on Cross Section of Castings from Heat B (Untreated) and Heat C (Zirconium Treated). Upper Specimen Untreated. Lower Specimen Zirconium Treated. × ⅛.

These were the first heats made in connection with this change in practice. I do not believe that Mr. Meissner would contend that the conditions were ideal so far as acid melting is concerned, but they do bring out certain information regarding recoveries which is instructive.

One observation which was made on the finished castings had to do with the sulphur content. The amount of zirconium added to these steels was in the neighborhood of 0.06 per cent or thereabouts, and there remained in the steel about 0.03 or 0.04 per cent of zirconium. Even this small amount of zirconium, however, was able to affect the action of sulphuric acid in making sulphur prints to a very marked degree.

The technique used in making the sulphur prints shown was developed by J. R. Vilella of the Union Carbide and Carbon Research Laboratories; he also prepared the photographs. In carrying out this printing, a very uniform

etching effect is obtained. The reagent employed was a three per cent aqueous solution of sulphuric acid.

Fig. 1 shows a cross-section of two castings from consecutive heats, B and C. The upper casting happens to have a few blow holes in it, which may not be particularly significant. Both of those sulphur prints were made under exactly the same conditions so far as time of exposure and strength of acid are concerned. You will notice a marked difference in the etching effect, although the amount of sulphur in those two steels is practically the same.

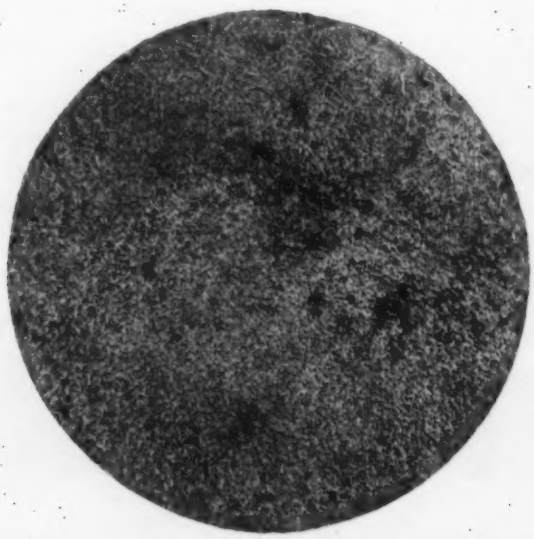


Fig. 2—Sulphur Print of 11-Inch Diameter Specimen Machined Eccentrically from 24-Inch Nickel Steel Ingot, Showing Portion of Ring of Sulphide Inclusions Which Occurred About Half Way Between Axis and Periphery of Ingot. Untreated Ingot.

All we can conclude from this particular picture, however, is that zirconium has formed a sufficient amount of zirconium sulphide to affect markedly the acid-etching characteristics. Obviously, the sulphide inclusions in this casting were attacked very slowly by three per cent sulphuric acid. The sulphur distribution in these two castings is fairly uniform, there being no marked segregation. Although representing a fairly high sulphur steel, the section is small (12 inches wide and about three inches maximum thickness).

Fig. 2 is a section taken from a 24-inch acid open-hearth ingot, but not from the heats referred to in the paper. This particular section was cut eccentrically, so as to include in an 11-inch diameter section a portion of the ring of sulphide inclusions which occurs about half way between the center and outside of the ingot. You can see there are pronounced black spots in this print, which are undoubtedly manganese sulphide.

Fig. 3 was made under the same conditions of practice and is a similar steel, a 3 per cent nickel steel to which a small amount of zirconium was added. While the sulphur content is practically identical with that of the preceding steel, the dark patches of manganese sulphide are absent and a

much more uniform distribution of sulphide inclusions in this particular zone is obtained.

These prints resemble somewhat certain sulphur prints we saw yesterday in connection with one of the other papers. The question of sulphur printing is one which is very readily carried out in the laboratory. It seems to me that this entire effect of zirconium is one which could be studied with a very small

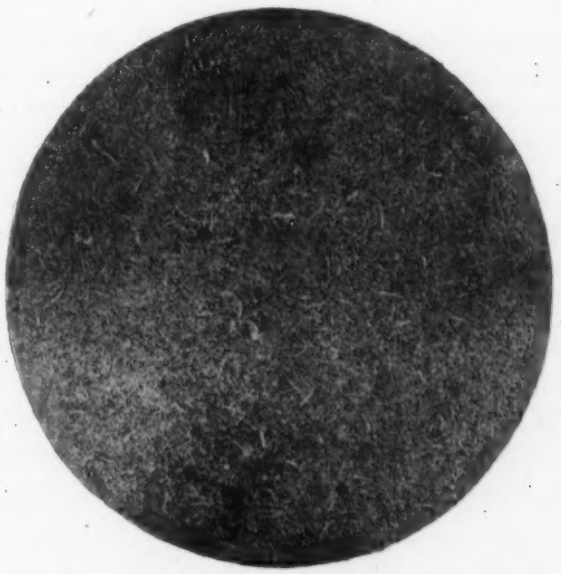


Fig. 3—Same as Fig. 2 Except from a Heat of Zirconium Treated Steel. Ring of Sulphide Inclusions is Absent. Zirconium Treated Ingot.

amount of difficulty in plant laboratories and results obtained for all sorts and conditions of practice. It is not important that zirconium sulphide etches light; the important point is the effect of zirconium on the size and distribution of sulphide inclusions rather than its collateral effect on etching rate. As to the effect of zirconium on the hot-rolling properties of high-sulphur steels, which is a subject of kindred interest, I would refer you to a paper* on this subject which I presented in 1924 before another society.

*Transactions, American Institute of Mining and Metallurgical Engineers, Vol. 70, p. 201.

THE APPLICATION OF SCIENCE TO THE STEEL INDUSTRY—SECTION VI

By Dr. W. H. Hatfield

Abstract

This, the last section of this paper, covers carbon and high speed tool steels. This discussion is accompanied by photomicrographs of each in variously treated conditions. The results are described of an investigation of table knives, razors, etc., to determine the characters of those possessing good cutting edges as compared to those with poor cutting edges. Some operations to be observed in the maufacture of cutlery are discussed.

Tool Steels and Cutlery

IN THIS section the author proposes to discuss a few questions arising in regard to the special class of steels coming under the above general heading.

Carbon Tool Steels

The carbon tool steels are an extremely interesting class, and for many purposes will repay careful study and knowledgeable treatment. In England, the higher grades of these steels are still manufactured by the crucible process and, as mentioned previously, this process lends itself admirably to producing homogeneous clean steel. Swedish iron is usually the base material, and undoubtedly crucible steel made from Swedish material has deservedly earned a very high reputation. Among those engaged in the manufacture and distribution of these steels, it is not unusual to hear them talk of that mysterious quality "body." When the Iron and Steel Institute visited Stockholm some eighteen months ago, a number of us had the pleasure of looking over the magnificent research laboratories of the Metallographic Institute, and

This paper is the sixth section of the third Edward De Mille Campbell Memorial Lecture presented before the tenth annual convention of the society, held in Philadelphia, October 8 to 12, 1928, which is being published in six sections in consecutive issues of Transactions. The author, Dr. W. H. Hatfield, is director of the Brown-Firth Research Laboratories, Sheffield, England.

during the visit, the author found an opportune moment to have the question put—"Will Professor Benedicks kindly define and explain to us the nature of 'body' in Swedish iron and steel?" The distinguished Professor, without any hesitation at all, answered in his most charming manner, "My mother, she told me not to tell such things."

Now as to whether and to what extent the best crucible steels of this class possess this mysterious characteristic, the author is unable to be dogmatic, but of this he is perfectly sure, that such steels, made from the purest materials under the best conditions, can be considered to possess the combination of qualities to which it is not surprising that time has given the mysterious appellation of "body." The behavior of a cutting tool, under service conditions, is determined by so many factors, many of which are only empirically controlled under what is termed "good practice", that it is difficult to determine, without an actual cutting test, what the exact merits of a particular steel are. Were the author asked to define the word "body", he would be a little more explicit perhaps than Professor Benedicks was on the memorable occasion previously referred to, and he would define the characteristics as "that outstanding reliability under strenuous conditions, which cannot be determined by simple direct tests, but can only be experienced under actual practical service conditions." It is naturally a foregone conclusion that an all-round high standard of technique as regards material, method of production and final heat treatment, must be reflected in the service obtained.

A careful consideration of the various details which must be controlled to obtain the maximum reliability and efficiency, entitle a really good tool steel to be looked upon as an undoubtedly superfine product. The materials must be carefully selected, the composition must be very well controlled, and the method of melting must include adequate "killing." The mold must be of proper design and the teeming done at the correct temperature and speed, with the care necessary to ensure that the feeder head is effective, and that the frozen ingot is free from pipe and gas holes, entrapped dirt or laps. The heating for forging should be done intelligently, since too rapid heating may cause clinks, exceeding a certain temperature may introduce the overheated condition, whilst insufficient heating or continuation of the hammering too

long or too heavily, may, during the working down stage, result in internal rupture of the material. A proper amount of work upon the material is also desirable. Add to this that the condition prior to the final hardening must be such that the final condition, as will be shown later, does not reflect after hardening an unsatisfactory condition of the material, and it will be agreed that when all this is achieved, the technician may be excused if he claims some mysterious merit of the material.

Generally speaking, these carbon tool steels range from 1.5 down to about 0.7 per cent carbon, and generally have a low manganese content. It is very desirable that, in heating for the forging operation, such material as just mentioned shall not be overheated, and the author considers that an initial temperature of 1920 to 2010 degrees Fahr. (1050 to 1100 degrees Cent.), according to carbon content, should not be exceeded, and that work should not be put upon the steel after the temperature has fallen to the neighborhood of 1292 degrees Fahr. (700 degrees Cent.).

Assuming that such steel has been carefully cast into small ingots of suitable shape with adequate feeder heads, and has been forged under proper conditions and given good technique in the final hardening and tempering operations, the most delicate and intricate tools of the highest characteristics and reliability can be produced. Particularly as regards the higher carbon steels, it is always well to anneal subsequent to the forging operation.

The microscope is invaluable in the study of such tool steels. While the understanding of the microstructure of purely iron-carbon steels is well understood, it cannot be said, that the same remark applies to complex alloys steels, but, nevertheless, the structures are always traceable to composition and past history, and give valuable information.

In no instance is this better illustrated than in the case of file steel. Fig. 18 is typical of the rolled condition in which the carbon in excess of the eutectoid composition is present in the form of membranes of carbide of iron, with the general structure elongated in the direction of rolling. Fig. 19 illustrates the same steel after air cooling from 1742 degrees Fahr. (950 degrees Cent.), and the appearance of the needles of carbide is distinctive. Hardened from this condition, the martensitic matrix would be rendered brittle by the dispersion through it of this form of the carbide. By

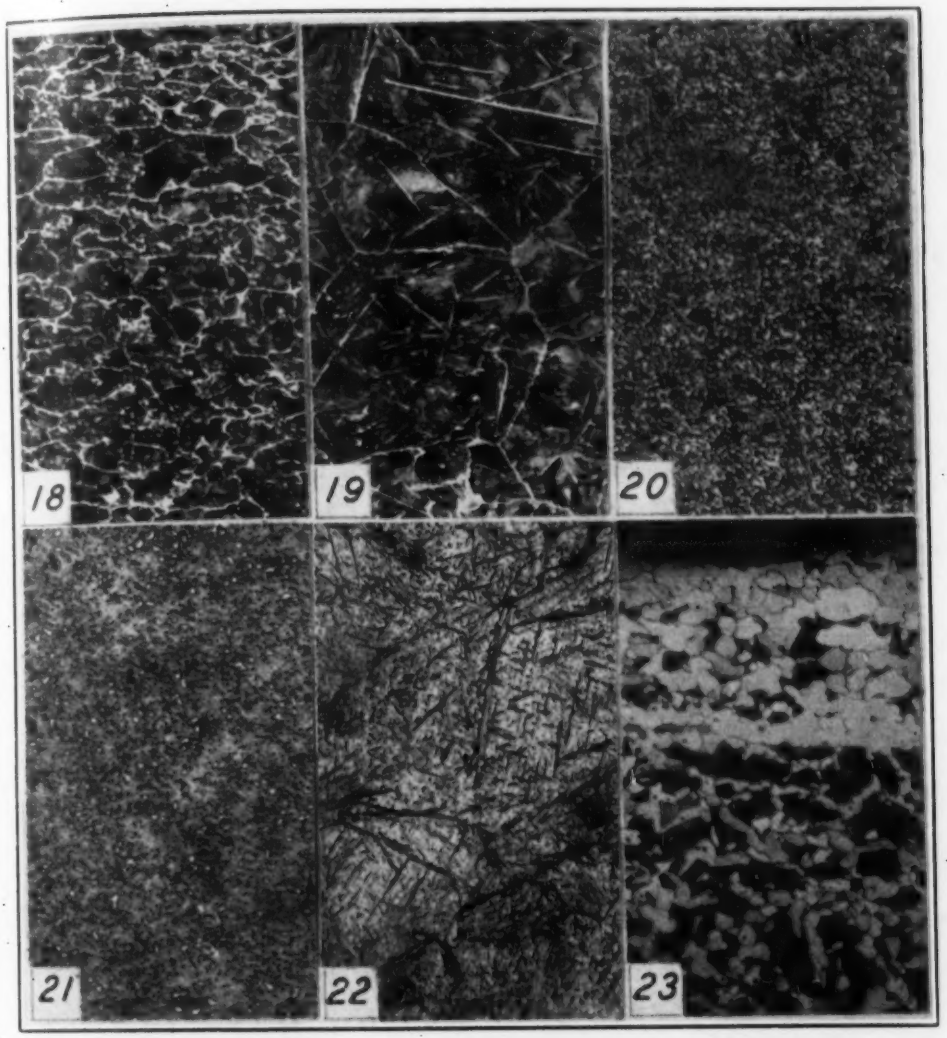


Fig. 18—Photomicrograph Typical of File Steel in the Rolled Condition. Fig. 19—Same Steel as Fig. 18 Heated to 1742 Degrees Fahr. (950 Degrees Cent.) and Air Cooled. Fig. 20—Photomicrograph of File Steel Annealed at 1364 Degrees Fahr. (740 Degrees Cent.) for 4 Hours After Forging. Fig. 21—File Steel Well Annealed and Hardened. Samples in Figs. 18, 19, 20 and 21 Etched in Picric Acid. × 200. Fig. 22—Photomicrograph of Carbon Steel Rock Drill Hardened from Too High a Temperature. Etched in Picric Acid. × 100. Fig. 23—Photomicrograph of the Bore of Hollow Carbon Steel Drill Showing Decarburization. Etched in Picric Acid. × 200.

a suitable low temperature annealing the structure shown in Fig. 20 is obtained, and hardening from this condition gives the correct structure for hardened file steel as shown in Fig. 21.

The same effect is common to other such high carbon steels, and Fig. 30 indicates the structure of a good safety razor blade, which obviously must have received the correct annealing treatment prior to being hardened.

Should carbon tool steel be hardened from excessively high temperatures, brittleness ensues and this effect may generally be detected as it is reflected in the microstructure, as shown in Fig. 22 which is taken from a sample of carbon rock drill steel. Again excessive time-temperature decarburizing effects are readily recognized as shown in Fig. 23 which illustrates a section through the bore of a hollow carbon steel drill.

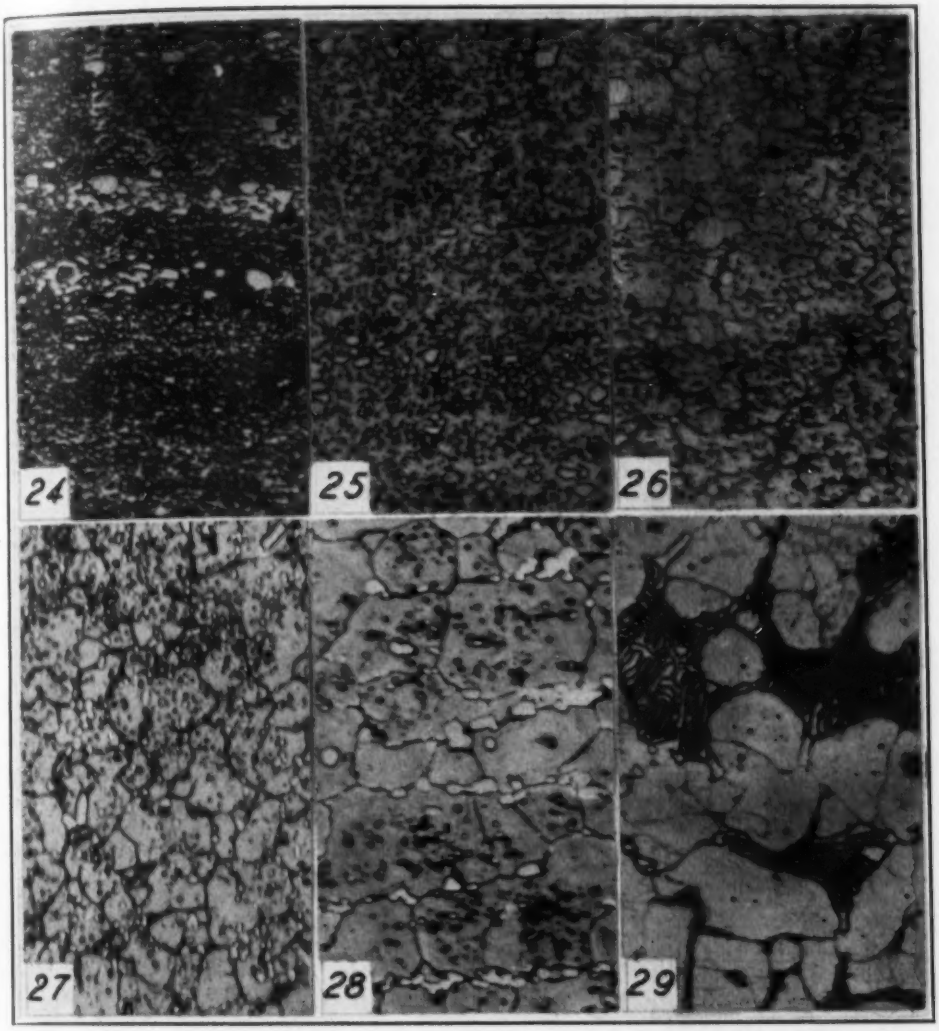
High Speed Steels

These steels are made over a wide range of composition and it may be of interest to you to have a few comments upon the subject. The metallography is complex and though many years have elapsed since the notable advances of Mushet and of Taylor and White, the subject is obviously still treated in an empirical fashion. The results of numerous valuable researches have been published over recent years, but we have still much to learn. All the care necessary in the production of carbon tool steel is required with additional steps to ensure homogeneity in composition, particularly in the steels richest in special elements.

The author has frequently been amused by the rival claims of different makers of such steels, claims out of all accord with actual facts. The steels vary in merit according to analysis, but each varies so much according to the treatment given, and the actual conditions under which comparative tests can be made are so complex, that such tests are only fair when all variables are eliminated. The optimum heat treatment for one steel is not necessarily the best for another.

The material which is being cut should be uniform as regards composition, condition and hardness. This is difficult to achieve in a bar of large diameter. The geometry of the tool should be carefully standardized; the speed, cut and feed be unified and under adequate control; while obviously each high speed steel should be given its optimum hardening and tempering treatment. Under such conditions accurate comparisons of merit can be made.

A few of the conclusions arrived at as a result of researches in the Brown-Firth research laboratories may be stated. The optimum carbon content lies between 0.55 per cent and 0.80 per cent. Below 0.55 per cent, there is difficulty in obtaining full cutting hardness. If above 0.80 per cent, the tools are not as



Photomicrographs of High Speed Steel. Fig. 24—Annealed. Fig. 25—Oil Quenched at 2102 Degrees Fahr. (1150 Degrees Cent.). Fig. 26—Oil Quenched from 2282 Degrees Fahr. (1250 Degrees Cent.). Fig. 27—Oil Quenched from 2372 Degrees Fahr. (1300 Degrees Cent.). Fig. 28—Quenched from 2507 Degrees Fahr. (1375 Degrees Cent.). Fig. 29—Air Cooled from 2552-2642 Degrees Fahr. (1400-1450 Degrees Cent.). All Specimens Etched with 2 Per Cent Nitric Acid in Alcohol. All Magnifications × 500.

resistant to shock, and furthermore, excessive wastage in manufacture occurs. Manganese, when added to excess, is a disability. Silicon may vary within normal limits without affecting the efficiency. The chromium generally should be within the limits 3.5 to 4.5 per cent. The tungsten may advantageously be replaced by molybdenum and cobalt. Long experience shows the advantage of 1.0 to 1.5 per cent of vanadium. If in the final stages of heating up, the heating is done slowly, the effect is detrimental from

two causes; decarburization and coarsening of grain. The optimum range of temperature for hardening is 2370 to 2450 degrees Fahr. (1300 to 1340 degrees Cent.), and the more rapid the cooling from the hardening temperature, the better will be the results with, of course, the exception that, under certain conditions, particularly of design, water quenching may produce cracking. Tempering to 1020 or 1110 degrees Fahr. (550 or 600 degrees Cent.), after hardening has definitely a beneficial effect upon the average life of the tool.

Hardness tests at normal temperatures give no indication of the possibilities of a cutting tool, and even if such quantitative tests are done at the higher temperatures, the figures are sometimes misleading. Properly controlled cutting tests as before described are the only reliable means of comparison. High speed steels have long been studied under the microscope, and in post mortems the microscope is invaluable. This is not because we completely understand the compositions of the constituents, but we are able to recognize their form, size and distribution and thereby gain much knowledge of the thermal history of the steel. This is well illustrated by photomicrographs taken from a steel containing 18 per cent tungsten, 3.5 per cent chromium and 1.5 per cent vanadium. Fig. 24 represents the structure of the steel in the annealed condition previous to the subsequent treatments. This particular photomicrograph is selected as containing the white constituent in a somewhat emphasized form. This constituent has been claimed to be carbide by some, tungstide by others; it most probably in this steel is a complex carbide of tungsten, iron and chromium. However, a knowledge of its constitution has not prevented the utility of the general structures obtained as evidenced by those depicted in Figs. 25, 26 and 27 which show the effect of gradually increasing the hardening temperature from 2100 to 2370 degrees Fahr. (1150 to 1300 degrees Cent.). The general effect appears to be the tending toward completion of the solid solution. Much higher temperatures as shown in Fig. 28 result in a considerable growth in grain size; 2370 to 2450 degrees Fahr. (1300 to 1340 degrees Cent.) has been established as the best hardening temperature. Fig. 29 is of interest as definitely indicating that at temperatures in the neighborhood of 2550 degrees Fahr. (1400 degrees Cent.)

the solidus has been passed during heating and that partial liquefaction had taken place.

Cutlery

One of the most interesting practical investigations for which our research laboratories have been responsible, arose through the suggestion, years ago, that knives made of stainless steel did not retain properly their cutting edge. So persistent was the suggestion that it was considered desirable to investigate the matter thoroughly once and for all. The ordinary steel blade, cleaned, as it is, either on a knife board or in a machine, has the surface abraded every time it is cleaned, and obviously this very action tends to keep the cutting edge sharp: whereas, knives made of

Table XXXVII

	Brinell Hardness Numbers
Good shear steel carving knives	504, 510, 525, 508
Good shear steel table blades	510, 525, 525, 530
Good ordinary steel carving knives	525, 520, 508
Good ordinary steel table knives	525, 508, 510, 512
Good cast steel pocket knives	560, 575, 560 580
Good cast steel razor	625, 625, 640
Good stainless carving knives	500/520
Good stainless table blades	500/550
Good stainless pocket knives	550/600
Good stainless razor	625
Bad-cutting stainless knives	380, 360, 400
Bad-cutting ordinary knives	400, 390, 385

Size of Brinell ball employed = 3/32-inch diameter.

stainless steel are simply washed, wiped and put away. It is, of course, quite clear that all cutting tools, including knives, must be sharpened from time to time.

However, since the investigation had to be made, it was thought that the best policy to pursue was to secure many examples of thoroughly "good cutting" carving knives, table knives, pocket knives, razors, etc., possessed by our several friends and acquaintances. Having, therefore, secured an excellent array of such examples, we proceeded to determine in a very accurate manner, the composition, hardness, and any other characteristics that might have a bearing upon the subject, including the geometry of the cutting edges.

The hardness is clearly a predominating factor in determin-

ing the life of the cutting edge and it was instructive to find that each type of knife had a characteristic range of hardness as shown in Table XXXVII.

It will be seen that good carving knives and table knives had a range of hardness which can be taken as 500 to 530 Brinell; pocket knives were definitely harder, having a Brinell hardness

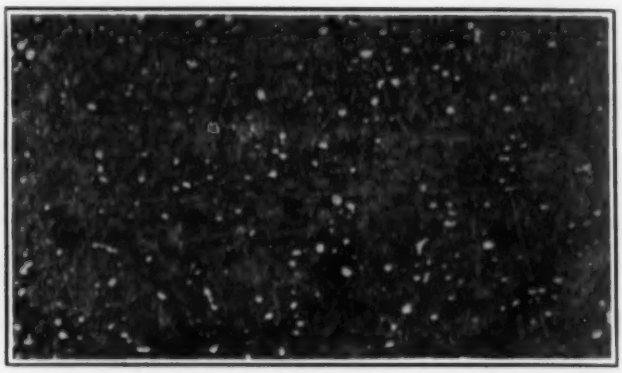


Fig. 30—Photomicrograph of an Excellent Razor Blade. Carbon 1.45 Per Cent, Manganese 0.13 Per Cent. Brinell Hardness 650. Specimen Etched with Picric Acid. × 200.

number of between 550 and 600, while razors clearly had a much greater hardness, the values ranging from 600 to 650. The consistency of these results is extremely interesting, having in mind that many of these knives had been manufactured as long as 40 to 50 years ago, and most probably it was the exception if any modern technical knowledge as regards temperatures had been applied.

Reference to the table will show that in the case of indifferently cutting knives, whether they were of stainless or of ordinary steel, the hardness was very much down. In commenting upon the hardness of the blade, one might observe that if a knife is too hard it obviously has not been sufficiently tempered; if it is too soft it is because, either (1) it has not been properly hardened, or (2) it has been over-tempered, always providing that the steel was of the correct composition. During the course of our studies the nature of the actual cutting edge was studied, and, generally speaking, it can be considered that this factor cancels out, since the characteristics were common to the blades made from both the ordinary and the stainless steels; this is illustrated in Fig. 32 and 33. While speaking of the quality of cutting edges, it is interesting to draw

Table XXXVIII

	Per Cent					
	Carbon	Manganese	Silicon	Sulphur	Phosphorus	Chromium
Shear steel table blades	0.80/0.95	0.05	0.07	0.01	0.01	....
Shear steel carvers	0.80/0.95	0.05	0.07	0.01	0.01	....
Cast steel table blades	0.80/0.95	0.22	0.12	0.03	0.01	....
Cast steel carvers	0.80/0.95	0.22	0.12	0.03	0.01	....
Ordinary steel table blades	0.45/0.50	0.50/0.80	0.12	0.05	0.05	....
Ordinary steel carvers	0.45/0.50	0.50/0.80	0.12	0.05	0.05	....
Cast steel pocket knives	0.80/1.0	0.22	0.12	0.03	0.01	....
Cast steel razors	1.4 /1.5	0.22	0.12	0.03	0.01	....
Stainless cutlery	0.30 (about)	0.30	0.12	0.02	0.02	13.0 (about)

attention to the characteristics in this respect, of two carbon steel safety razor blades, see Figs. 34 and 35, one of which gave excellent results and the other very poor results. As regards the geometry, i.e., the angles formed by the cutting edges, these also cancel out, since the variations are common to the same type of blade irrespective of the type of steel employed.

In the old days, and even now to some extent, the cutlery trade has been content to buy steels of certain tempers. This is largely due to the fact that the high-class Sheffield cutlery upon which Sheffield's reputation has been built, was manufactured from steel produced by the cementation or carburization of Swedish wrought iron. The blister bars produced in this way were then either piled into faggots and hammered into bars to produce shear steel, or they were broken up and melted in the crucible to produce crucible cast steel. It was left to the skill of the men who examined the broken blister bars to determine to which temper a particular bar belonged. In dealing with this matter I would not make a single criticism. The men who were responsible for the classification of the bars into the different tempers did their work extremely well. When, however, we come to a scientific consideration of cutlery, it is not enough to know the temper. We must know exactly how much carbon, manganese, silicon, sulphur, phosphorus or other element is present in the steel. We therefore, for instance, in our investigation dissolved up these excellent specimens of knives after performing our other tests upon them, and were thus enabled, by chemical analysis, to see exactly what

their composition had been. The results of our investigations will be found in Table XXXVIII.

It will be seen that good shear steel cutlery has an extremely low sulphur and phosphorus, a low manganese content, and, what is most important, a carbon content ranging from 0.80 to 0.95 per cent. With regard to cutlery made from crucible steel produced by the melting of the blister bar, I would observe that analysis alone reveals little difference except a slight increase in the manganese and sulphur content. It can, however, readily be determined whether a knife is made from shear or cast steel, as will be shown later.

It will be seen that the pocket knives made from cast steel are a little higher in carbon than the table cutlery. When we come to the razor we find that the carbon has been increased up to and above 1.40 per cent.

It will be noticed in the table that we have included the analyses of knives made from ordinary steel. This ordinary steel is manufactured by either the Siemens or the Bessemer process, and attention might be drawn to the fact that the carbon is very much lower and the manganese very much higher.

If the analysis of stainless steel is examined in the table it will be seen that, curiously, the carbon is as low as 0.30 per cent while we have present 13 per cent of chromium. It is the presence of this chromium which renders possible the production of the stainless knife. The application of a high chromium steel for this purpose is to be credited to our research laboratories during the period when its work was supervised by my friend Mr. Brearley, and the development of stainless steel knives will undoubtedly rank as one of the most important metallurgical advances of recent years.

One very interesting, though perhaps not as useful, side of our investigations consisted of polishing the various blades and then etching them with an 8 per cent solution of copper ammonium chloride. This reagent will be found to deposit copper on the blade, but this deposit should be removed while wet, and the knife rubbed with cotton wool and washed in a stream of water. The result is that in shear steel rather curious and pretty patterns are obtained, as shown in Fig. 31.

If these figures are studied it will be seen that by this means we get a concrete and definite knowledge of the manner of manu-

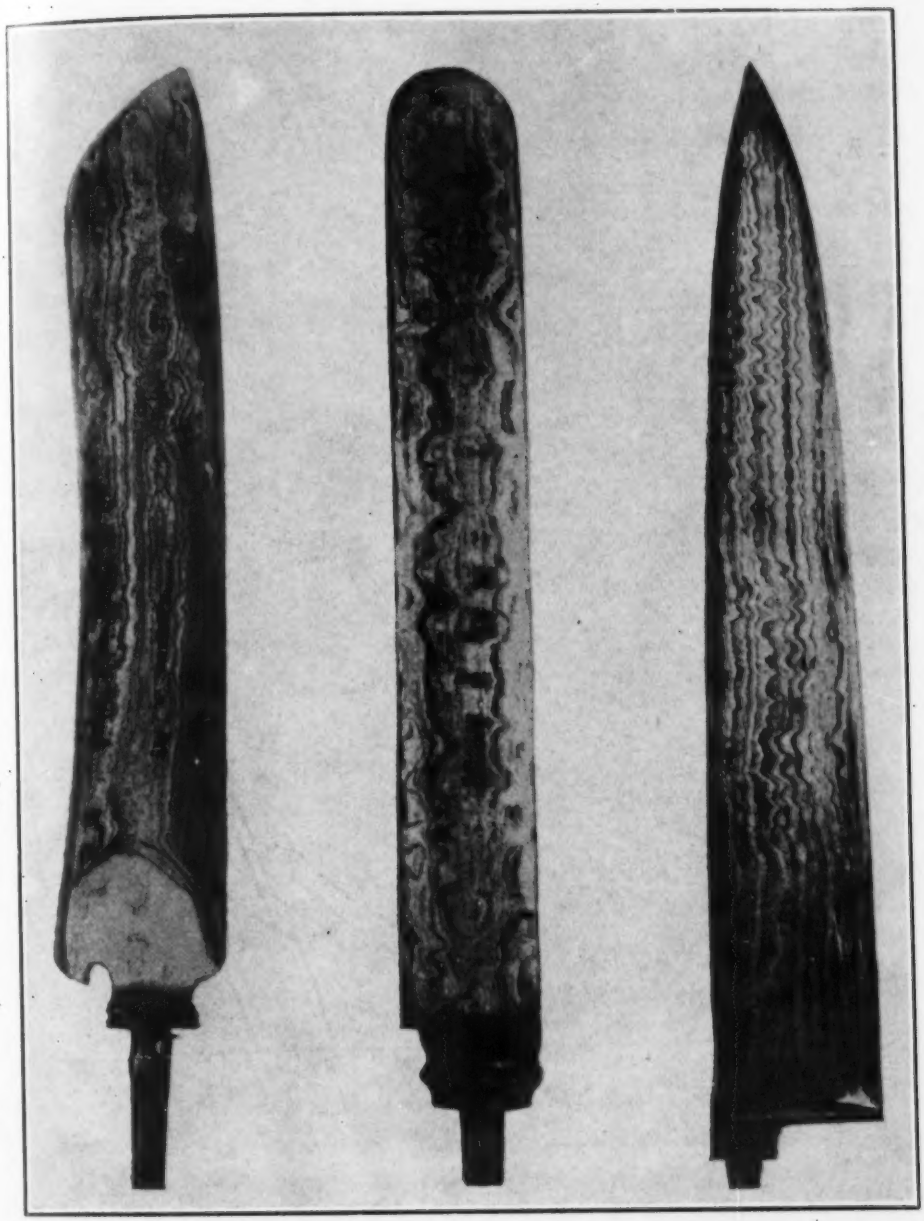
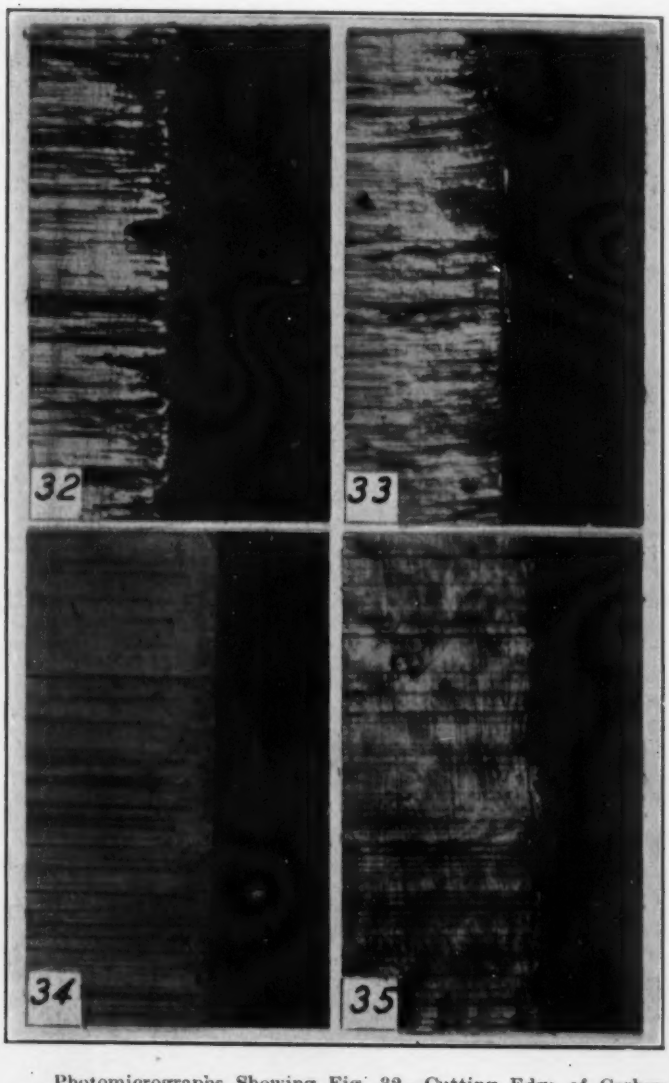


Fig. 31—Photograph Showing the Macrostructure of Table Knives.

facture of the steel from which the blade was made. These lines clearly indicate that the material is shear steel, and a careful consideration of them will even give one such precise information that one can determine whether the material employed was single shear or double shear, and, incidentally, how many bars were heated in each faggot. However, it is not proposed to say much on



Photomicrographs Showing Fig. 32—Cutting Edge of Carbon Steel Knife. Fig. 33—Cutting Edge of Stainless Knife. Fig. 34—Cutting Edge of Good Safety Razor Blade. Fig. 35—Cutting Edge of Poor Safety Razor Blade. All Magnifications × 100.

this subject except to point out that there is a ready method by which a shear steel blade can be distinguished from an ordinary cast steel blade, since the cast steel blade, under this treatment, does not give such a pattern if made from sound steel.

One might here, perhaps, make a few observations on what we consider to be matters where trouble is largely experienced in the manufacture of stainless knives. I have already pointed out that stainless steel hardens when cooled in the air from a high tem-

perature, and it is therefore clear that before "flying out" cold, the rough blades should be suitably annealed at 1380 to 1400 degrees Fahr. (750 to 760 degrees Cent.). If the blades are suitably softened, as they may readily be, it is best to fly them out cold. If the blades are not annealed, then they must be flyed out at a dull red, heating to which temperature will do much to soften the blades. It is unnecessary to mention the importance, in the first place, of not getting the steel too hot during the forging operation, and, in the second place, not putting work on it after it has become too cold. With regard to the smithing operation, i.e., straightening before hardening, it will be clear that if the blades have not been annealed but are smithed after cooling from a high temperature, some of them will probably be cracked in this process. Perhaps one of the most important points to watch is that excessive scaling is not permitted when the steel is exposed to high temperatures. If the scaling is excessive, difficulty is encountered in producing a perfect surface by grinding. If any of this scale, or oxide of iron, is left in the final blade, it forms a seat of corrosion and becomes a very pronounced defect in the course of time.

As regards the method of putting on the name and identification marks, the only process to employ is that of etching. Etching can be done with ease and gives a very good result, while stamping in the name, results in scale being pushed into the letters in such a way it cannot be removed, and instances have been know where the scale so introduced into the blade has formed the seat of corrosion.

In conclusion, it might be pointed out that blades should not be ground in such a way that they become heated. If they do, discoloration is the result, and this discoloration is indicative of a condition of the blade in which it does not perfectly resist staining influences. After this discoloration has been produced it may be removed by further polishing, but even so, the blade is not as perfectly stain resisting as if it had been ground with care and had not been allowed to become unduly heated in that operation.

For a stainless blade to be perfectly satisfactory the steel must be of the correct composition. The next point is that it must have been satisfactorily hardened from a temperature of 1740 to 1830 degrees Fahr. (950 to 1000 degrees Cent.) and have required and received the necessary tempering at about 360 to 390 degrees

Fahr. (180 to 200 degrees Cent.) to give it the temper required for service. It is possible to so harden stainless cutlery that the hardening operation will leave it with the correct temper, thus leading to the omission of the tempering operation. Knives, however, which have been hardened in that way are not quite as stain-resisting as when the blades have been properly hardened to begin with, and then had the tempering done subsequently. There is no doubt also that the finish on the blades is worth studying. A blade properly hardened and tempered, which has been ground, i.e., properly "bottomed" and polished to a really good finish, is better than the same blade when indifferently finished. All properly manufactured stainless blades will resist total immersion in 1.20 sp. gr. nitric acid. Hence, this nitric acid test is the one which should be employed for discriminating between stainless and other cutlery. It does not need to be said that stainless blades, if properly manufactured, will resist the acids, etc., encountered in ordinary domestic use.

In this short section I have endeavored to place before you the results of what have been very interesting investigations, and it can be summed up briefly in the following words that:—

(1) The cutting properties of cutlery are largely determined by the success with which the exact hardness required is obtained.

(2) This hardness should be periodically determined during manufacture in a quantitative manner on the lines indicated.

(3) It has been clearly shown that the requisite hardness for perfect cutting properties can be obtained in stainless blades of all kinds, when properly hardened and tempered.

(4) The respective characteristics of shear, best cast, and ordinary steel, have been considered, and it has been shown that whereas shear and high carbon cast steel should be hardened from 1400 to 1440 degrees Fahr. (760 to 780 degrees Cent.) and the ordinary steel a little higher, stainless steel should be hardened from temperatures of 1740 to 1830 degrees Fahr. (950 to 1000 degrees Cent.).

(5) That in the case of stainless steel knives, the hardening should be effective in the first place and that tempering should be done at 360 degrees Fahr. (180 degrees Cent.).

THE PHYSICAL CHEMISTRY OF RIMMED STEEL

By J. E. Carlin

Abstract

The first part of this article deals briefly with the early method of manufacture of low carbon steels by the basic open-hearth process, and the start and development of the rimming process. The second part explains rimmed steel from a physical chemistry standpoint.

Historical

THE terms "rimmed" "open" and "effervescent" are applied to low carbon steels manufactured by the basic open-hearth and Bessemer processes to which little deoxidation is applied with the exception of manganese, either in the furnace, the ladle or the molds. The terms "open" and "effervescent" were prompted by the fact that after the steel is in the molds, the evolution of gas proceeds for a considerable length of time and in appearance is directly opposite to killed or deoxidized steels which lie dead and flat. In the early days of steel manufacture, the main product of the basic open-hearth process was low carbon steel. Before the advent of ferrosilicon and aluminum, steel was deoxidized by ferromanganese only, which in the ordinary range of 0.30 to 0.80 per cent, is far from complete. In order to hold the steel in the molds it was necessary to cover the top of the mold with a heavy cast iron cap or cover, and to hold this in place by means of a steel bar placed across the cap and fitting tightly into two lugs, one on each side. When the ingot had solidified sufficiently the cap was removed. The violent evolution of gas during solidification caused irregular freezing on the mold walls, which resulted in enormous slivers in the subsequent blooms and billets. Blow holes near the surface resulted in small slivers, seams and laps when the ingot was scaled in the soaking pits and rolled in the blooming mills.

Rimmed steel was the result of attempts to improve the quality by controlling the evolution of gas and directing its escape in

The author, J. E. Carlin, a member of the society, is on the metallurgical staff of the Union Drawn Steel Company at Beaver Falls, Pa. Manuscript received December 3, 1928.

an orderly manner through the top only of the ingot. This necessitated doing away with the cap and pouring the ingots short, since the steel came up a few inches before solidifying. The improvement was slow and the action of the steel in the molds could never be predicted with certainty. Even today it is a question whether or not a heat will rim properly. It was found that in a large number of cases the steel would grow a few inches and then start to solidify around the edges, leaving the center in a turbulent state owing to the evolution of gas. As the steel cooled, the solidification which started around the edges progressed uniformly from each side toward the center, leaving at all times, however, a sufficient opening for the free evolution of gas. The term "rim" takes its name from the progressive solidification inwards of the top of the ingot. If, for unknown reasons, conditions had been just right, the steel would free itself of gas at the time the rimming had reached the center from all sides. In other cases, it was found that instead of starting to solidify shortly after the finish of the pour, the steel continued to grow and to prevent it from going over the top a cap had to be used. It also happened that although the rimming action had started early it progressed too fast, and the top of the ingot had solidified before the gas had been completely evolved. The pressure of the gas under the thin skin was sometimes great enough to burst through and throw molten steel for a considerable distance. This dangerous occurrence was prevented in a large measure by placing a heavy cap on the top of all rimmed heats after the top had frozen.

The importance of carbon and manganese content, as well as temperature was early recognized, but it was difficult to regulate the temperature and the manganese content seemed to be fixed for economic reasons. The importance of carbon content was obvious as it was found that steels with a carbon content up to 0.15 per cent would rim much better than steels with a higher carbon content. The rimming process today is applied successfully only to steels under 0.15 per cent carbon. Although the use of aluminum and ferrosilicon has changed the manufacture of low carbon steels, the rimming process is still used to a large extent. Low carbon steels contain, comparatively, a much greater amount of dissolved ferrous oxide than high carbon steels. During the rimming process much of this ferrous oxide, together with man-

ganese oxide comes out of solution and being lighter than the metal rise to the surface as a slag. A comparatively large amount of silicon or aluminum is required to deoxidize low carbon steels and the deoxidation products are retained to a large extent. In the rimming process no silicon or aluminum are added and the carbon monoxide and ferrous oxide are removed by natural means. The rimming process, if correctly carried out gives an ingot with a thick skin of dense metal surrounding a core of spongy metal which contains most of the impurities. If the skin is thick enough and the ultimate product is sheet steel, wire, etc., the spongy center is not harmful and the rimmed steels are preferred to low carbon killed steels due to maximum softness and freedom from non-metallic inclusions in the surface.

Physical Chemistry of the Process

The fundamental reaction for carbon removal in the open-hearth furnace is

$$C + FeO \rightleftarrows Fe + CO - 36540 \text{ Calories}$$

When an open-hearth heat is tapped it is assumed that the carbon, ferrous oxide, iron and carbon monoxide in solution, are practically in a state of equilibrium, because, regardless of whether equilibrium exists between slag and metal, any increase of ferrous oxide concentration in the metal results in a corresponding decrease of carbon content at the same temperature. The equilibrium expression for the above reaction is

$$\frac{(C) \times (FeO)}{(CO) \times (Fe)} = K$$

The dilution of the system due to iron formed is negligible and iron may be omitted. The expression would then become

$$(C) \times (FeO) = K\ (CO)$$

which means that at any definite temperature at equilibrium, the product of the concentrations of carbon and ferrous oxide is directly proportional to the concentration of carbon monoxide. It has been proven[1] that at equilibrium the product of the concentrations of carbon and ferrous oxide is a constant. The logical

[1] C. H. Herty, Jr., Bureau of Mines Bulletin No. 34.

reasoning, therefore is that the concentration of carbon monoxide is also a constant, depending entirely upon temperature and independent of both carbon and ferrous oxide content.

When any open-hearth steel heat is tapped, ferromanganese is added either in the furnace, the ladle or in both. At tapping time as assumed above, the four components C, FeO, CO and Fe are in solution and at equilibrium. When manganese in the form of FeMn is added it is immediately subject to both of the following reactions:

$$(1)\quad Mn + FeO \rightleftarrows MnO + Fe + 25200 \text{ Calories}$$
$$(2)\quad Mn + CO \rightleftarrows MnO + C + 61800 \text{ Calories}$$

It would appear that the manganese added had a choice between these two reactions. The deciding factors, however, which determine the extent and rate of any reaction are temperature, concentration of the reacting substances and their affinities. An inaccurate, but sometimes workable measure of the relative affinity of elements and compounds for each other, is in the amount of heat evolved in their respective reactions. It could be assumed that as more heat is liberated in reaction (2) manganese would have a greater affinity for CO than for FeO. In low carbon steels, however, the concentration of FeO greatly exceeds that of CO and for this reason, manganese will react more intensely and completely with FeO up to its equilibrium concentration than it will with CO. The point to be taken is that manganese will react with both CO and FeO but in low carbon steels the FeO concentration is great enough to overcome the greater affinity of manganese for CO. The result being a partial removal of FeO and a slight decrease in CO content. Regardless, then, of carbon content, everything else being equal open-hearth heats tapped at the same temperature contain the same amount of CO in solution. After the manganese addition low carbon heats contain more CO as the manganese partially deoxidizes but does not degasify.

When the heat is poured it can be assumed that there are now in solution and in equilibrium C, Mn, CO, FeO and MnO. The steel first solidifies on and near the walls of the mold. The effect of the actual solidification is to set free the dissolved CO for although the solubility of the gas decreases with increasing temperature it is comparatively insoluble in the solid metal. The effect of the removal of some CO from solution is to renew the reaction

$$C + FeO \rightarrow CO + Fe$$

as it attempts to regain equilibrium. This results in a further evolution of CO through the oxidation of an equivalent amount of carbon. The obvious paradox of an endothermic reaction proceeding with increasing speed at falling temperatures is explained by the fact that an equilibrium may be displaced by increasing the concentration, or by removing one of the reacting substances. In this case the removal of CO by the solidification of a small amount of metal allows the reaction to proceed to the right with the formation and evolution of more CO. The reaction increases in speed as more metal solidifies due to a more complete removal of the reaction product (CO).

Up to a certain point, the freezing point of a solution varies inversely with the concentration of the solute. If steel is considered a solution of carbon in iron, the freezing point will be depressed as the percentage of carbon is raised. In other words high carbon steels remain liquid at a temperature at which low carbon steels solidify. In rimmed steels a rapid solidification on the mold walls is imperative as the gas is thus prevented from escaping anywhere but through the top of the ingot. A rapid solidification also insures a thick skin of dense metal, and any entrapped gas or blowholes will be found in the interior where they are practically harmless. The two most important factors in the manufacture of rimmed steel, therefore, are carbon content and temperature, for low carbon content and low temperature both promote rapid solidification.

A summary of the points which have practical value are as follows:

(1) The amount of carbon monoxide present in a steel before the addition of manganese is dependent on temperature only, and is independent of carbon and ferrous oxide content. (2) After the manganese addition high carbon steels contain less gas (CO) than low carbon steels and for this reason it is impossible to rim high carbon steels. (3) The evolution of gas (CO) depends upon actual solidification of the metal. (4) Carbon content and temperature are the most important factors in the manufacture of rimmed steel.

A STUDY OF BURNING AND OVERHEATING OF STEEL—PART I

By W. E. Jominy

Abstract

Steel which is being heated for forging may become overheated resulting in burning. Inasmuch as there is a natural tendency to forge at high temperatures it is well to know the conditions affecting burning and the underlying causes of this behavior of steel. This paper covers a research on this subject.

The effect of oxidizing and reducing atmosphere on steels heated for forging were studied and the results given.

IT is quite well known that if a piece of steel is heated too hot and forged, the steel will fracture, or may even fly to pieces during the forging. This steel is said to be burned. In some cases, although the steel is burned, it does not appear to fracture at the forge, but proves to be very brittle and fails at a later time, showing a fracture that is extremely coarse, the fracture being around the grains and not through them. This type of failure occurring in service often results in serious consequences. Burned steel may usually be recognized from the characteristic appearance of the fracture of the failed steel. When the burning is slight, however, it is necessary to make a metallographic examination. Examined in this way, burned steel shows a separation of grains in certain areas leaving voids between the grains.

Since there is a natural tendency to forge at as high a temperature as possible due to the greater ease of forging at these higher temperatures, it is important to know the limiting temperatures in order to avoid burning and the conditions surrounding it.

A review of the scientific literature on the subject of burning and overheating of steel showed that relatively little experimental work had been done. Some of the earlier researches undertaken

This paper constitutes part of an investigation carried out for the American Gas Association on the subject of forging, at the University of Michigan.

The author, W. E. Jominy, member of the society, is associated with the Engineering Research department of the University of Michigan, Ann Arbor, Michigan. Manuscript received January 7, 1929.

to explain the phenomenon of burning, were, however, quite significant. Among the more important investigations were those carried on by Osmond,[1] Stead,[2] Stansfield,[3] and Brearly.[4] The evidence presented by Stansfield and Stead indicates that heating above the solidus line in the iron-carbon diagram will cause steel to burn. The position of the solidus line in the iron-carbon diagram is, however, still open to considerable difference of opinion. For instance, if 1.1 per cent carbon steel is considered, and it is decided to determine the highest temperature at which this steel may be heated without burning, it may be found by reference to the iron-carbon diagram of Carpenter and Keeling,[5] that the solidus line for this steel is at 2372 degrees Fahr. and this temperature, therefore, must not be exceeded if burning is to be avoided. If the position of the solidus line in the iron-carbon diagram according to Gutowsky[6] is accepted, then this steel may be expected to burn if heated above 2140 degrees Fahr. There is here an indicated difference in the burning point of about 230 degrees Fahr.

In studying the reported work of both Carpenter and Keeling, and of Gutowsky, it seems surprising that neither reported trouble with their thermocouples. In carrying on the present investigation, one of the major troubles encountered was with the thermocouples and a very considerable expense was involved in using new couples, since the deterioration of the couples was very rapid. The writer was interested in learning of the trouble reported by Tritton and Hanson[7] which was very similar to his own.

The present research was initiated with a view to studying the conditions affecting burning of steel and, if possible, to learn the underlying causes of this behavior of steel. In particular it was desired to learn the effect of various atmospheres on the burn-

[1]Osmond, *Bulletin*, d'Societe d'Encouragement pour l'Industrie Nationale, Vol. 10, 1895, p. 776.

[2]Stead, *Journal*, Iron and Steel Institute, Vol. 53, No. 1, 1898, p. 145. *Transactions*, Northeast Coast Institute of Engineers and Shipbuilders, Vol. 29, 1913, p. 142.

[3]Stansfield, *Journal*, Iron and Steel Institute, Vol. 64, No. 2, 1903, p. 433.

[4]Brearly, *Forging, Stamping and Heat Treating*, Vol. 12, 1926, p. 290.

[5]Carpenter and Keeling, *Journal*, Iron and Steel Institute, Vol. 65, 1904, p. 224.

[6]Gutowsky, *Metallurgie*, Vol. 6, 1909, p. 731.

[7]Fred S. Tritton and D. Hanson, "Iron and Oxygen." *Journal*, Iron and Steel Institute, Vol. 110, No. 2, 1924, p. 90.

ing of steel and the temperature at which certain common steels burn. Another objective of this study was to determine whether steels might be burned entirely on the inside of the section so that careful visual examination of the outside would reveal no defect. So large a number of tests were made in the course of this work, that it is impractical to present all the data in one paper. For convenience, therefore, the tests have been classified as: (1) heating tests without subsequent forging; and (2) heating tests with subsequent forging. This paper is limited to a presentation of the first class of tests. It is intended to describe first the furnaces used in the entire research and to discuss the manner of measuring temperatures, and then to give an account of the results bearing on the behavior of the steels when heated to high temperatures in strongly reducing atmospheres, in oxidizing atmospheres, and in vacua.

Furnaces

Three heating furnaces were used, including an electric carbon-resistor furnace and two gas furnaces. One of the gas furnaces was a muffle furnace and the other had the gas burning inside the heating chamber. Only a few preliminary heats were made in the electric furnace. The heating chamber in the electric furnace was 6 inches square by 10 inches deep. The steel sample was placed on a firebrick 2¼ inches above the bottom of the furnace and the platinum, platinum-rhodium thermocouple was placed in the furnace through the top so that the bottom of the thermocouple rested within ½ inch of the piece. The thermocouple was protected by a refractory closed-end protecting tube.

One of the gas furnaces had a vertical heating chamber 10 inches in diameter by 20 inches deep. The gas-air mixture was introduced at diagonally opposite sides at the bottom of the heating chamber. The steel to be heated was placed on a small platform of firebrick which was at the center of the heating chamber and stood about 5 inches above the bottom of it. In this position the steel was not in the path of the flame, but was surrounded by combustion gases. The temperature of the sample was measured by a thermocouple placed in the furnace through the side.

The muffle furnace which was employed for some of the tests consisted of a D-shaped carborundum muffle 5 inches wide at the bottom, 14 inches long and 4½ inches high. This muffle was

heated by two gas burners placed diagonally opposite and staggered. The sample to be tested was placed on a firebrick 1¼ inches thick and the temperature measured by a platinum, platinum-rhodium thermocouple placed through the back of the furnace. This muffle was also connected with a gas line and an air line so that an atmosphere of raw gas, air or combinations of these could be maintained in the muffle. Temperature measurements made at various points in the muffle at 2400 degrees Fahr. showed a maximum variation in the muffle temperature of 8 degrees Fahr., except within two inches of the door.

Thermocouples

Platinum, platinum-rhodium thermocouples were used. For the preliminary tests the usual thermocouple mounting was employed. This consisted of a small two-hole insulating tube which in turn was mounted in a closed-end protecting tube. The protecting tube was always within half an inch of the steel sample being tested and this separated the bead of the thermocouple from the steel sample. At the high temperatures used it was obviously impossible to have the unprotected thermocouple bead held against the steel surface. This meant that the furnace temperature was being measured near the steel sample and that a heat-insulating material, the refractory tube, separated the bead of the thermocouple from the steel sample.

To remedy this possible source of error, a special thermocouple mounting, as shown in Fig. 1, was devised. In using this equipment, the large tube was pushed against the steel specimen, so that the platinum cap was in contact with the steel. Since there were two caps, the thermocouple bead was separated from the steel surface by a small air gap of about 1/64 inches, and 0.04 inches of platinum. This method of measuring temperature was used throughout all the tests after the preliminary work. The platinum caps seemed to offer as good protection against the chemical action of furnace atmosphere as did the closed-end porcelain tubes first used. The life of the thermocouple, however, was rather short, and one bead of the thermocouple seldom lasted more than ten tests. The thermocouples were tested for accuracy before each day's run. Homogeneity tests of the thermocouple wires were also made, and bad spots on the wires near the bead were often found,

more often on the pure platinum wire than on the platinum-rhodium alloy. Originally when the couples became contaminated they were annealed and treated with concentrated hydrochloric acid. Such long annealing was required, however, to correct the thermocouples that it was finally found to be more satisfactory to cut off the contaminated wire and reweld. Usually not more than

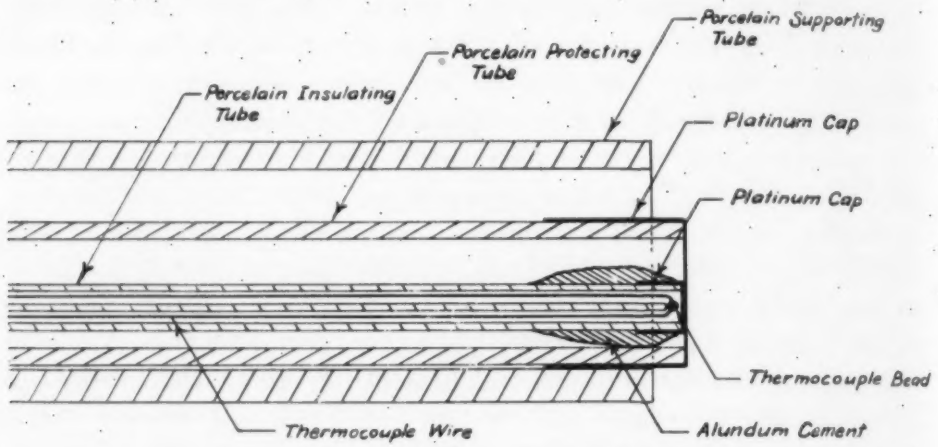


Fig. 1—Special Thermocouple Mounting.

two inches of wire from the bead was contaminated. As to the cause of the contamination of the thermocouples the writer is inclined to agree with Tritton and Hanson[8] who reported similar trouble. In their opinion "the couple contamination appeared to be caused by the iron vapor contaminating the couple protector." It has been observed, that, when the thermocouple protecting tube is brought close to the steel at temperatures around 2500 degrees Fahr., the protecting tube becomes spattered with brown spots. These spots increase in number with time of exposure, and increased temperature, and after a short time the tube end near the steel becomes covered with a brown crust which presumably is an iron compound.

Experimental Procedure

One set of tests was made by heating steel samples to forging temperatures and then allowing them to cool, usually in air, to room temperature. A record set of tests was made by heating these steel samples to forging heats and forging. The steels tested are listed in Table I.

[8]Loc. cit.

Table I
Chemical Analyses of Steels Tested

Steel No.	Carbon Per Cent	Manganese Per Cent	Sulphur Per Cent	Phosphorus Per Cent
1	0.22	0.51	0.047	0.014
2	0.47	0.64	0.020	0.022
3	0.51	0.65	0.030	0.025
4	0.55	0.55	0.013	0.018
5	0.85	0.30	0.014	0.011
6	0.93	0.30	0.016	0.004
7	1.10	0.28	0.015	0.004

The first series of tests was made on Steel No. 1 (0.22 per cent carbon). Sections about ½ inch long were cut from a ¾-inch round bar of this steel and heated in a carbon-resistor furnace to 2200, 2300, 2400, 2500, 2600 and 2725 degrees Fahr. and cooled in air. These samples were then cut in half and one-half the sample polished and examined under the microscope and the other half normalized by heating to 1650 degrees Fahr. and cooled in the furnace, after which they were polished and examined under the microscope.

Microscopic examination of the samples showed no evidences of burning, except in the sample heated to 2725 degrees Fahr. In this sample there were large voids between the crystals and there was also evidence of decided decarburization around the burned areas. The other half of this burned specimen was normalized but was not restored to its normal structure by the normalizing treatment. The voids, of course, were the same as before, but the grain boundaries of the original gamma iron grains were still in evidence.

All the samples heated to 2400 degrees Fahr. and higher showed decarburization. The higher the heating, the more pronounced the decarburization. Since the burned areas were decarburized, tests were made to determine how the steel would behave if heated at the burning temperature in an atmosphere which was not decarburizing.

Accordingly a sample of steel No. 1 was placed in a graphite crucible provided with a close fitting graphite plug for a cover. This crucible containing the piece was placed in the furnace after the furnace temperature had reached 2725 degrees Fahr. and held at this temperature for twenty minutes. The crucible was then removed and cooled in the air.

After cooling, the piece was cut in half, and one-half nor-

malized at 1650 degrees Fahr., while the other half was polished without further treatment. Examination of the microstructure of the piece which had been untreated after heating to 2725 degrees Fahr. showed no indication of burning. The corners of the piece had been carburized and partially melted but there was no evidence of voids anywhere.

The other half of the section which had been normalized also showed no evidence of burning, but its microstructure showed that normalizing had not restored it to normal structure, since the boundaries of the large original grains of austenite still persisted.

Action of Reducing and Oxidizing Atmospheres

After these tests, further studies were made on this steel, using a gas-fired furnace in which the atmosphere was controllable. Accordingly, a piece of No. 1 steel was heated in a gas furnace for fifteen minutes at 2700 degrees Fahr. in a strongly reducing atmosphere produced by firing the furnace with a large excess of gas. Microscopic examination of this sample showed no evidence whatever of burning. See Fig. 2.

Another sample of this same steel was then heated to 2600 degrees Fahr. for fifteen minutes in a strongly oxidizing atmosphere produced by firing the furnace with a large excess of air. The microstructure of this sample showed it to be burned. See Fig. 3.

Still another sample of this steel was heated to 2500 degrees Fahr. for fifteen minutes in a strongly oxidizing atmosphere. This steel was not burned. After normalizing this steel at 1650 degrees Fahr. it showed no evidence of original gamma iron grain boundaries. It thus appeared that the burning point for this steel was somewhere between 2500 and 2600 degrees Fahr. in a strongly oxidizing atmosphere. Such a temperature is below the lowest solidus temperature given in any iron-carbon diagram for this steel. Furthermore, these results indicated that, in a strongly reducing atmosphere, the steel did not burn at any temperature up to 2725 degrees Fahr. at which the steel began to melt at the corners of the sample. This latter temperature is above the highest solidus temperature given in any iron-carbon diagram for this steel. It appears, therefore, that the atmosphere in which the steel is heated is an important factor in the so-called burning of the steel.

Tests were then made on Steel No. 4 (0.55 per cent carbon).

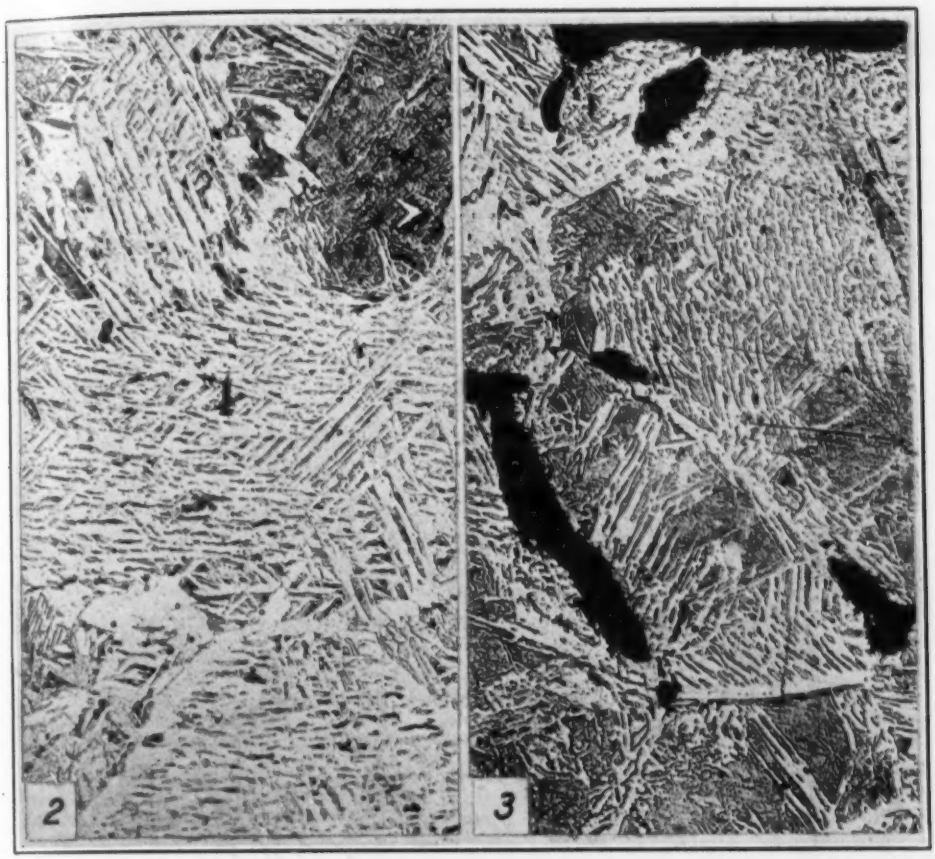


Fig. 2—Photomicrograph of 0.22 Per Cent Carbon Steel Heated in a Strongly Reducing Atmosphere to 2700 Degrees Fahr. Note that this Sample is not Burned.

Fig. 3—Photomicrograph of 0.22 Per Cent Carbon Steel Heated to 2600 Degrees Fahr. in a Turbulent Strongly Oxidizing Atmosphere. This Sample is Burned. All Specimens Etched with 4 Per Cent Nitric Acid in Alcohol. × 100.

Accordingly samples were heated separately in a gas-fired furnace for fifteen minutes, at 2200, 2300, 2400 and 2500 degrees Fahr. in a strongly oxidizing atmosphere. This steel did not burn at 2200, 2300 or 2400 degrees Fahr., but burned at 2500 degrees Fahr.

The same tests were applied to Steel No. 5 (0.85 per cent carbon steel) and to Steel No. 7 (1.10 per cent carbon steel). Steel No. 5 was found to burn at 2425 degrees Fahr. in a strongly oxidizing atmosphere, but not at 2300 degrees Fahr. The burning temperature of Steel No. 7 was found to be somewhere between 2300 and 2400 degrees Fahr. in a strongly oxidizing atmosphere.

Since the observations of the effect of atmosphere on burning are entirely new, at least so far as it has been possible to discover

from the scientific literature or elsewhere, confirmatory tests were made.

A piece of Steel No. 2 (0.47 per cent carbon) 2 inches long by 2¼ inches square was placed in a graphite crucible. Strips of graphite were placed above the steel but not in contact with it and the crucible was then closed with a graphite plug. The furnace was heated to 2650 degrees Fahr. and the crucible placed in the furnace. This cooled the furnace and it was brought back to 2650 degrees Fahr. after which it was held at this temperature for 30 minutes. When cooled it was found that the steel had begun to melt, but there was no evidence of burning, although there was a slight carburization.

From the same bar of Steel No. 2 another sample, two inches long, was cut. This sample was placed in the gas furnace, which had been brought to 2625 degrees Fahr. with a reducing atmosphere. It was held in the furnace for twenty minutes after which the gas was turned off and the piece allowed to cool at 2400 degrees Fahr. in the furnace. It was then removed and cooled in the air. The gas analysis of this reducing atmosphere showed 6.4 per cent carbon monoxide, 10.0 per cent carbon dioxide and no free oxygen. After this treatment the steel was sectioned and examined under the microscope and found to contain a small burned area. The sample was also somewhat decarburized.

Still another section two inches long was cut from the bar of 0.47 per cent carbon steel. This sample was placed in the gas furnace, which had previously been brought to 2650 degrees Fahr. with a strongly oxidizing atmosphere. It was held there for twenty minutes and then removed from the furnace. The steel sample was full of large holes and oxide right through to the center of the section, the worst case of burning the writer has ever seen.

Figs. 4 and 5 are facsimile photographs of the steel samples after treatments in strongly reducing and in strongly oxidizing atmospheres (Steel No. 2). These photographs give an idea of the relative difference in these samples.

From these three tests it appears that Steel No. 2 may be heated to 2650 degrees Fahr. without burning if the atmosphere is sufficiently reducing, whereas, in a turbulent oxidizing atmosphere at the same temperature, the steel will be badly burned.

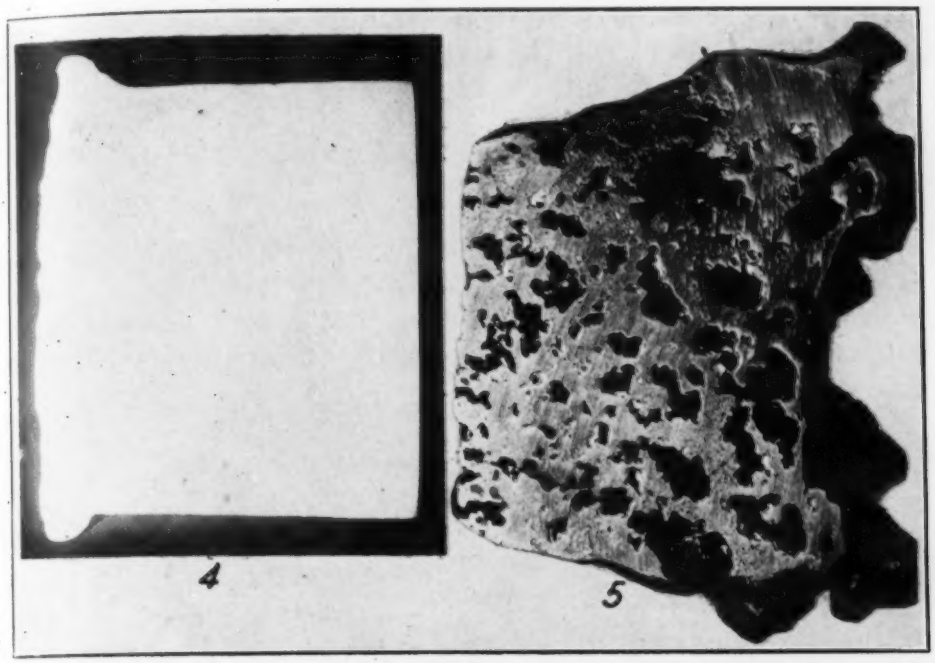


Fig. 4—Photograph of Sample of 0.47 Per Cent Carbon Steel Heated to 2650 Degrees Fahr. for 30 Minutes in a Strongly Reducing Atmosphere.
Fig. 5—Photograph of Sample of 0.47 Per Cent Carbon Steel Heated to 2650 Degrees Fahr. for 20 Minutes in a Turbulent Strongly Oxidizing Atmosphere.

To confirm further the resistance to burning in a strongly reducing atmosphere, a study of Steel No. 3 (0.51 per cent carbon) was made. Two sections 1¼ inches long were cut from the ¾-inch round bar of this steel. One of these sections was placed in a graphite crucible with a graphite plug as a cover and the other section was placed in a clay crucible covered with an alundum cover. These two crucibles with the covers were placed in the gas muffle furnace, after it had been heated to 2400 degrees Fahr. The furnace was then brought up to 2650 degrees Fahr., and was held at this temperature for twenty minutes, at the end of which time the furnace was shut down and the door opened. In ten minutes' time the temperature had dropped to 2350 degrees Fahr. and the crucibles were removed and allowed to cool in the air. When the samples were removed from the crucibles, it was noted that the sample heated in the graphite crucible had been partially melted. The sample which had been heated in the clay crucible had not melted and appeared sound although it had slumped slightly. Upon cutting the sections in half and polishing, it became evident that the unmelted portion of the steel heated in

the reducing atmosphere, in the graphite crucible, gave no indication of burning, whereas, the steel heated in the oxidizing atmosphere, in the clay crucible, was burned on the inside of the section.

Effect of Reducing Atmosphere

The reason the steel samples were not burned in the reducing atmosphere, although they were partially melted, seems to be due to the carburizing action of the strongly reducing atmosphere. As these samples are heated they become carburized, thus forming an outer shell of high carbon steel around the core of lower carbon steel. As the temperature of the piece is increased, it finally reaches the melting point of the high carbon steel and, as this begins to melt, it absorbs some heat. At this higher temperature it carburizes more rapidly and the process keeps ahead of the melting of the high carbon shell so that another shell is formed as rapidly as the previous shell melts so that the steel melts entirely on the outside. If the steel can be made to melt entirely on the outside, there are apparently no voids or separated crystal boundaries, and, since the atmosphere is carburizing, there can be no internal oxidation. When these phenomena are avoided, there is no burning.

If this theory is correct, then, a strongly reducing atmosphere will not protect a high carbon steel from burning, since in this case there can be no low carbon core. Having a high carbon core the steel will melt on the inside as well as the out, which melting will produce voids and burning. Experiments have proven this to be the case; that is, a Steel No. 6 (0.93 per cent carbon) is not appreciably protected from burning by a strongly reducing atmosphere.

Effect of Oxidizing Atmosphere

After this conclusion has been reached regarding the behavior of steel in reducing atmospheres, it was decided to investigate more completely the behavior of steel in oxidizing atmospheres at these high temperatures. Inasmuch as Steel No. 1 burned at a furnace temperature over 100 degrees Fahr. lower in an oxidizing atmosphere than in a mildly reducing one, a number of tests were made to throw some light on this phenomenon.

A sample of Steel No. 2 (0.47 per cent carbon) was heated in the gas furnace in an oxidizing atmosphere to 2500 degrees Fahr. To do this the furnace was first brought to equilibrium at 2500 degrees Fahr. then the sample placed in the furnace against the

platinum cap of the thermocouple tube. The temperature dropped and then increased to 2500 degrees Fahr. and continued rising. The furnace was turned down to prevent further increase of temperature and the piece brought to equilibrium at 2515 degrees Fahr. After the piece was removed, the temperature of the furnace dropped to 2450 degrees Fahr. although no further adjustment of the furnace valves had been made, indicating that the furnace atmosphere had been cooler than the steel. The steel was found to be burned.

This test was repeated exactly as described above excepting that a reducing atmosphere was used instead of an oxidizing one. The steel came to the temperature of the furnace at 2500 degrees Fahr. and the temperature remained there for twelve minutes, at the end of which time the steel was removed. After the removal of the piece, the temperature of the furnace remained constant. The steel was not burned.

Another test was made in which the temperature of the furnace was brought to 2510 degrees Fahr. with an oxidizing atmosphere and a firebrick the same dimensions of the steel placed in the furnace against the thermocouple. After the furnace temperature was constant, the brick was moved away from the thermocouple and a piece of Steel No. 2 placed against the thermocouple. The temperature at once dropped and then came back to 2510 degrees Fahr. and then began to rise higher than the previous temperature of the furnace atmosphere. Although the fuel adjustment was left unchanged, the temperature of the piece continued to rise until it reached 2587 degrees Fahr. which is 77 degrees Fahr. above the furnace temperature. It remained at this temperature until the steel was removed from the furnace. As soon as the steel was removed from the furnace the temperature dropped to the former temperature of 2510 degrees Fahr. showing that the temperature of the furnace atmosphere had been less than that of the steel. To show further that the increased temperature was due to a reaction of the steel, the firebrick was now placed against the thermocouple. The temperature of the thermocouple now remained the same and did not increase for five minutes, at the end of which time the furnace was turned off and the experiment considered completed. The steel sample heated in this test was burned.

To show that this increased temperature was due to the oxygen in the furnace atmosphere a piece of Steel No. 2 was placed in a

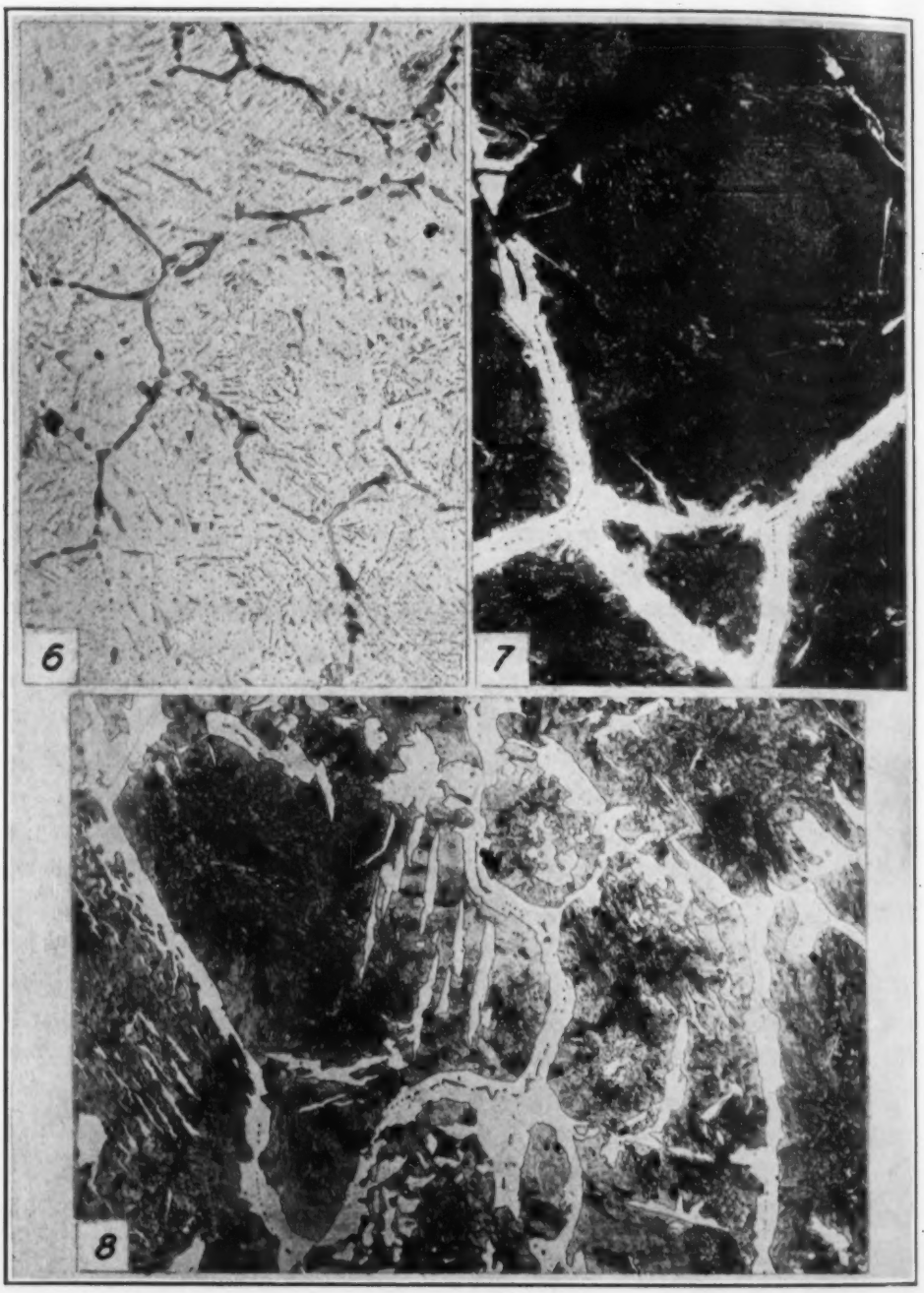


Fig. 6—Photomicrograph of 0.22 Per Cent Carbon Steel Heated to 2450 Degrees Fahr. in an Oxidizing Atmosphere for a Long Period of Time.

Fig. 7—Photomicrograph of 0.51 Per Cent Carbon Steel Heated to 2650 Degrees Fahr. in an Oxidizing Atmosphere. Note the Nonmetallic Inclusions in the Grain Boundaries.

Fig. 8—Photomicrograph of 0.47 Per Cent Carbon Steel Heated to 2714 Degrees Fahr. in a Vacuum. Note the Nonmetallic Inclusions in the Grain Boundaries. All Specimens Etched with 4 Per Cent Nitric Acid in Alcohol. × 100.

clay crucible which was placed in the gas furnace. A small porcelain tube was cemented into the bottom of this crucible so that air could be passed into the crucible. A cover was luted onto the crucible and a small hole drilled into the cover so that the gas could escape. The furnace was then brought to 2550 degrees Fahr. and the air turned on to flow at the rate of 0.35 cubic feet per minute. The temperature of the piece began to rise, although the incoming air should have cooled it, and the furnace was turned down. The temperature of the piece, however, rose to 2600 degrees Fahr. before it could be stopped. When this piece was removed from the furnace a hole was found to be burned into it as though it had been cut with an acetylene torch. This hole was no doubt melted out by the shaft of air as it came from the tube and first struck the steel. Further tests were made in the muffle furnace, but these will be omitted to save space, since the results were all confirmatory.

What has been proved here is that, under certain conditions, steel in the solid state will actually become hotter than the furnace atmosphere in which it is placed. It is the writer's belief that this is the result of the reaction

$$3\ Fe + 2\ O_2 \rightarrow Fe_3O_4 + 270{,}800 \text{ calories}$$

When the rate of oxidation is slow, the excess heat is carried off in the surrounding atmosphere as fast as it is liberated. When, however, the rate of oxidation is rapid, as it is at these high temperatures, the heat is liberated faster than it is dissipated, and consequently the temperature of the piece rises considerably above that of the furnace atmosphere. When the oxide is formed from carbon dioxide heat is absorbed according to the reaction.

$$3\ Fe + 4\ CO_2 \rightleftarrows Fe_3O_4 + 4\ CO - 13{,}000 \text{ calories}$$

It follows that free oxygen must be present to cause the steel to become hotter than the furnace atmosphere. A large number of tests have been made where this phenomenon was observed and in no case did it occur below a temperature of 2475 degrees Fahr. As a general rule the reaction started at about 2500 degrees Fahr.

Heating in Vacuum

To study still further the action of atmospheres on burning it was thought desirable to heat one of the steels in vacuum. As no vacuum furnace was available at the University of Michigan, Dr. Louis Jordan of the U. S. Bureau of Standards heated samples of

Steel No. 2 in the vacuum furnace at the Bureau of Standards.

Three samples 0.7 inches square by two inches long were used in this test. These were heated in an Arsem vacuum furnace, one to a temperature of 2714 degrees Fahr., one to a temperature of 2732 degrees Fahr. and one to a temperature of 2750 degrees Fahr. in a vacuum of 5 millimeters of mercury. The temperature in this case was measured with an optical pyrometer.

The samples heated to 2732 and 2750 degrees Fahr. were too completely melted to be of any value in this study. The sample that had been heated to 2714 degrees Fahr. had not been heated enough to change its outside dimensions and was examined under the microscope. This sample showed a tendency to form voids in straight lines as though the steel melted and ran down. Examination at 100 diameters showed typical voids at the grain boundaries and in this respect the structure resembled the burned samples heated in oxidizing atmospheres.

In nearly all the samples examined, burning had been the result of partial fusion of the steel. There have been some exceptions, however. Fig. 6 shows the structure of Steel No. 1 which is undoubtedly very much weakened by reason of the iron oxide between the crystals. This steel was heated several times to 2450 degrees Fahr. held at heat from one to three hours, and then cooled in the air after each heating. This treatment led to the formation of oxide between the steel crystals as shown in the Fig. The penetration of the iron oxide into the grain boundaries is not uniform and extends into the sample only about $\frac{1}{8}$ inch from the surface.

It is unlikely that there was any fusion in this sample since its maximum temperature was only 2450 degrees Fahr. The writer is at a loss to explain how iron oxide can diffuse into this sample in view of its recently reported very slight solubility in iron. Furthermore, why should the oxide follow the crystal boundaries?

A possible explanation of this phenomenon is that the oxidizing furnace gases in which this steel was heated diffused into the steel between the crystals and oxidized the surfaces of the grains of austenite. This explanation presupposes that there is room between the crystals of austenite to allow a slow penetration of oxygen or carbon dioxide. Whether this supposition is justified must be determined by future experimental work.

In some of the samples which have been burned by partial fusion of the steel, iron oxides or other nonmetallic inclusions have

been found in the grain boundaries. This, however, is due to an entirely different cause from that described above and shown in photomicrograph, Fig. 6. Fig. 7 is an area of steel No. 3. This steel was heated in a clay crucible for 30 minutes in the muffle furnace at 2650 degrees Fahr. Fig. 8 presents an area of Steel No. 2 heated in a vacuum to 2714 degrees Fahr.

Both of these samples contain nonmetallic inclusions strung along the grain boundaries in areas where there are no voids. That the inclusions in this case have not been caused by oxidation at the surface is indicated by the fact that they are present in the grain boundaries of the steel heated in a vacuum. Apparently these inclusions are rejected to the grain boundaries by the crystals just before fusion takes place in the grain boundary.

Summary

To summarize briefly the results of the experimental work, the following conclusions may be stated:

1. The word "burning" as used in the industry is not a good word, since it implies chemical reaction of the steel with oxygen. This reaction is not necessary to produce so-called "burning." In practically every case which the writer has observed there have been voids around the crystal boundaries of "burned" metal. In this work steel was always considered burned when it contained voids which followed the crystal boundaries.

2. A reducing atmosphere seems to protect plain carbon steels having a carbon content of 0.55 per cent or less from burning. If the atmosphere is reducing enough to produce carburization, these steels will not burn. This is not true of high carbon steels containing 0.85 per cent carbon or more.

3. A turbulent, strongly oxidizing atmosphere will cause these lower carbon steels to burn at relatively low furnace temperatures; a 0.47 per cent carbon steel will burn at 2500 degrees Fahr. in such an atmosphere.

4. When heated in a turbulent, strongly oxidizing atmosphere to furnace temperatures of 2500 degrees Fahr. or higher, plain carbon steels actually become hotter than the furnace atmosphere in which they are heated.

5. The burning of steel has been found to occur in a vacuum, that is, voids between crystals are produced, probably due to the fact that the crystal boundaries melt out at these temperatures.

Recommended Practice Committee Releases

NICKEL CAST IRON

By Dr. Paul D. Merica*

Nickel cast iron usually contains from 0.25 to 5.0 per cent nickel in addition to the usual ranges of silicon and carbon.

Nickel is added to gray iron for the purpose of obtaining improved and more uniform structure, hardness, wear resistance, strength in readily machinable castings, and to widen the limits of hardness and to promote machinability. Nickel, however, must be used with care and discretion if economic benefit is to be derived from it. There is no universal formula for nickel iron, but each problem requires its own special prescription.

Effect of Nickel on Gray Iron—Nickel in gray iron depends primarily on the fact that it reduces chilling tendency without appreciably coarsening the graphite structure (or grain) or without lowering the strength. At the same time nickel hardens the pearlite matrix of the iron. The effects of nickel in gray iron may be enumerated as follows:

1. It reduces chill and the local or general formation of free iron carbide. In this respect it acts just as does silicon and indeed may replace or supplement silicon in rendering castings gray. Two parts of nickel are generally equal to about one of silicon in reducing chill, although with low silicon content (under 1.0 per cent) the relation is nearer three or four to one.

2. In addition to reducing and eliminating free carbide, nickel reduces pearlitic combined carbon mildly, particularly up to about 1.5 per cent nickel. With larger amounts of nickel the pearlitic combined carbon is slowly increased again up to about 0.80 per cent, the eutectoid amount.

3. It is generally without marked effect on the graphite structure or grain of iron. It does not coarsen fine grained or hard iron appreciably and may mildly refine the structure in medium hard or medium grained iron.

4. Nickel enters into solid solution in the ferrite or pearlite matrix of the iron and changes it progressively into sorbite and even martensite, with over 5.0 per cent nickel.

5. Nickel therefore hardens progressively the pearlitic matrix of iron. See Fig. 1.

6. It is generally without marked effect on the strength of iron. It does not, however, weaken strong or hard iron appreciably, and generally strengthens iron mildly when added in amounts over 1.5 per cent. See Fig. 2.

7. The effect of nickel in amounts up to 5.0 per cent on such properties as melting point, fluidity, contraction and shrinkage, viz., those properties

*Assistant Manager, Development and Research Department, International Nickel Co., New York.

of iron which determine the ease of producing castings, is practically negligible.

Machinability and Structure—The three features noted above, viz., Nos. 1, 3, and 4 constitute the useful effects of nickel in iron. It reduces chill by promoting primary graphitization, but at the same time hardens and strengthens the pearlitic matrix and does not coarsen the graphite. The value of nickel in iron is primarily related to its machinability. The

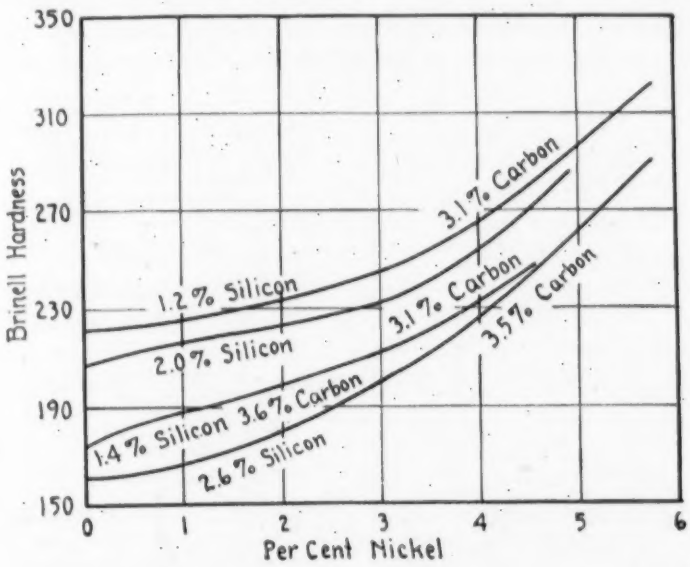


Fig. 1—The effect of Nickel on the Brinell Hardness of Gray Iron (Arbitration Bars).

use of nickel permits of the production of castings, which in comparison with plain iron castings are:

1. either harder, stronger, and of better structure with equal machinability,
2. or are more readily machinable for the same hardness, strength, and structure.

The machinability of gray iron depends on its structure, and primarily upon the absence of free carbide, or of white or chilled iron, locally or throughout the casting. Iron which is completely gray in structure and contains no free carbide or cementite is readily machinable, but the presence of free carbide impairs machinability and in substantial amounts renders machining practically impossible.

Free carbides or white iron first make their appearance in the thin sections, the corners, edges, and surfaces of castings, which cool more quickly than the heavier sections or body. The machinability of a casting is therefore limited generally by the chilling of its thinnest portions and the occurrence there of free carbides. For every individual casting a grade of iron must be used which will not chill in its thinnest section.

The grade or iron, its chilling tendency, and consequently its ma-

chinability are controlled by varying the silicon content. As this is raised, chilling tendency is diminished, and machinability improved; but the structure becomes more open and the hardness and strength decrease. As silicon is lowered, conversely, the iron is rendered harder and stronger, but at the expense of machinability.

The Step Bar Test—These all-important relations between silicon content, structure, and machinability are well recognized generally by foundrymen, but they are not, however, adequately taken into consideration in the current arbitration bar test of gray iron, which gives a measure of the strength of any iron, but not of its *suitability for the production of machin-*

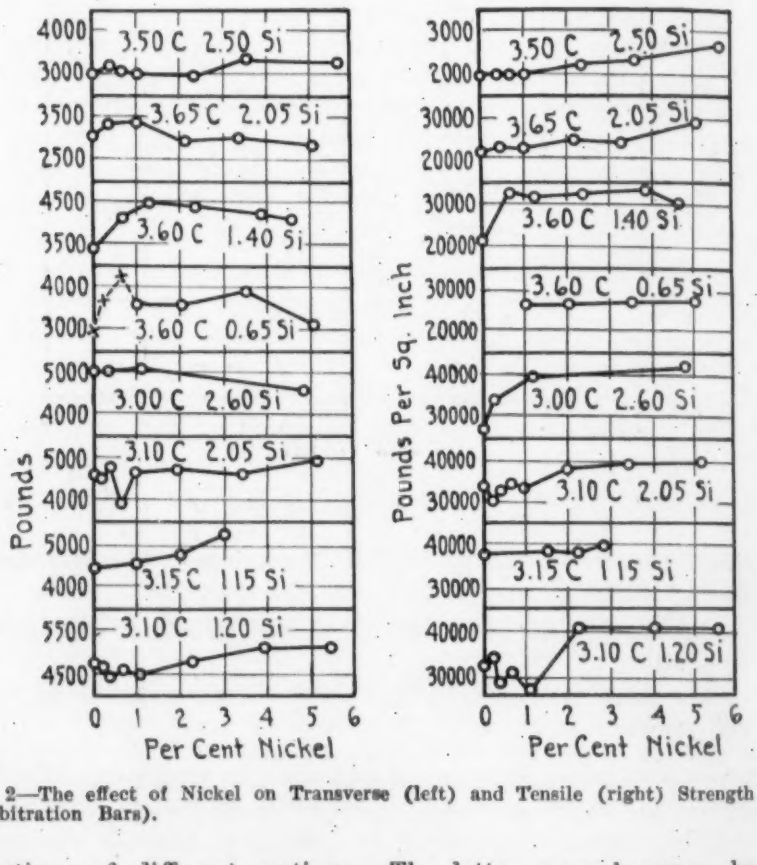


Fig. 2—The effect of Nickel on Transverse (left) and Tensile (right) Strength of Gray Iron (Arbitration Bars).

able castings of different sections. The latter may, however, be tested conveniently by means of the step bar test casting as shown in Fig. 3. The different steps of this test bar are fractured and examined, and the hardness and combined carbon content determined on the different sections. These results are considered in conjunction with the transverse and tensile tests of arbitration bars.

If carried out on a series of plain irons of varied silicon content, the step bar test demonstrates that as the silicon content is decreased, the hardness and strength of heavy sections increase, but at the expense of increasing chill and decreasing machinability of lighter sections. A series

of such step bar tests is shown in Fig. 3. These illustrate the fact that it is not possible in plain gray iron to secure Brinell hardness values (on arbitration bars) much over 200, nor strengths much over 4,000 pounds (transverse) and 35,000 pounds (tension) without using such a low silicon or carbon content as to render light section castings mottled or white and unmachinable. These are the natural limits of machinable plain gray iron.

It is, however, possible to exceed these limits by the proper use of nickel. In Fig. 4 are shown similarly several step bar tests of iron to which nickel has been added in progressively increasing amounts. As the nickel is increased the hardness and strength of the heavier sections increase slightly, whereas at the same time the lighter sections become progressively grayer and more machinable.

Nickel Replaces Silicon—The advantages of low silicon iron for castings are well recognized. The iron is of finer structure, contains more combined carbon, and is harder and stronger. Nickel supplements silicon in reducing chill, and thus permits of the use of hard and strong low silicon iron in castings which without the nickel would be unmachinable. It is indeed only by the use of nickel in irons of lower* silicon content that the full economic benefit of nickel additions to iron can be secured, and simple additions of nickel to iron are not generally economically justified.

The general principle to follow, therefore, in using nickel in iron is to lower the normal silicon content of the iron by 0.25 to 1.0 per cent and to add the equivalent amount of nickel to reduce chill and restore machinability. When the general statement is made that nickel refines the grain of iron and increases its strength, reference is made to iron in which nickel has been substituted for silicon in proper and equivalent measure.

Pearlitic Structure—It is well recognized that in order to secure full pearlitic structure in castings cooled under normal conditions, it is necessary to use a low silicon iron, and that when the silicon content lies much above 1.50 per cent the combined carbon content of castings even of medium section is likely to decrease below the eutectoid amount required for full pearlitic structure. Full pearlitic structure is desirable in castings because of its better wear resistance and hardness, and principally because of the more uniform structure of pearlitic castings in heavy and light sections, and consequent freedom from internal casting stresses.

Low silicon nickel iron is pearlitic in structure as would be expected from its silicon content, and in addition is readily machinable because of the chill reducing effect of the nickel.

Uses of Nickel Iron

The economic use of nickel in gray iron is governed by the principles discussed above. Their application to individual foundry problems cannot be summarized as simply and concisely, and are best illustrated by actual typical examples of current foundry practice. These illustrations will also

*Lower than the normal silicon content for the particular casting in question. Of course there are other adjustments of composition which are at least partially equivalent to reduction of silicon content; e. g., reduction of total carbon content, increase of sulphur content, etc.

serve to emphasize again the fact that it is generally necessary when using nickel in iron for any particular casting, to make at the same time other adjustments in the composition of the iron, in order to secure the maximum economic value of the nickel addition.

Increasing Hardness and Wear Resistance of Machinable Castings—This is a problem frequently presented to the foundry in connection with a wide variety of types of castings; generally in connection with a casting which in plain iron is sufficiently machinable, but not hard enough. In such cases from 0.75 to 2.0 per cent nickel should be added (or even more for maximum results) and at the same time the silicon content of the iron should be lowered correspondingly. In this manner the hardness may be

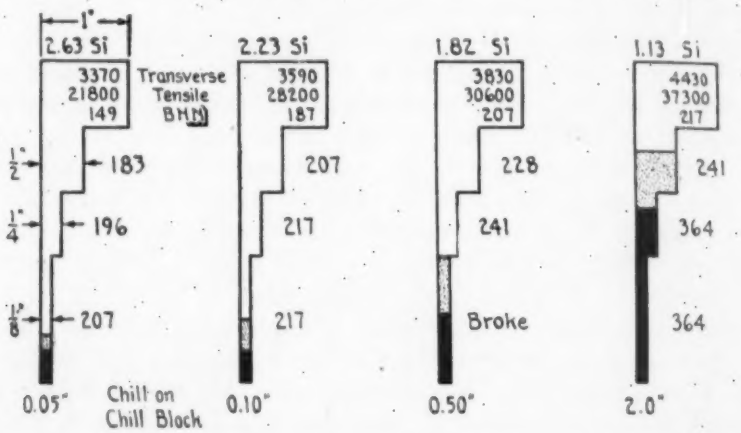


Fig. 3—Hardening Gray Iron by decreasing the silicon content. The Step Bar Test.

increased by 20 to 50 points Brinell, due first, to the direct hardening effect of the nickel on the iron matrix (pearlite), and second, to the increased combined carbon (pearlitic structure) caused by lowering the silicon content.

1. *Automobile Cylinder Block*—A block with a bore of 3⅛ inches made in plain iron, gave about 150 to 160 Brinell hardness in the bore which showed a partly ferritic structure. By lowering the silicon content and adding about 1.5 per cent nickel the Brinell hardness was increased to 180 to 190, and the structure changed to substantially complete pearlitic. No change was noticed in machinability, but in consequence of the better structure and increased hardness the wear resistance of the cylinder bore was about doubled, as indicated by service tests.

COMPOSITION OF CYLINDER BLOCKS

	Alloy Iron Per Cent	Plain Iron Per Cent
Nickel	1.25–1.50	
Silicon	1.80–2.00	2.20–2.40
Carbon	3.20–3.40	3.20–3.40
Manganese	0.40–0.50	0.40–0.50
Sulphur, under	0.10	0.10
Phosphorus, under	0.20	0.20

2. *Tractor Cylinder Bushings*—Bushings of plain iron ¾ to ⅞ inch

in section, showed about 160 Brinell hardness in the bore. By lowering the silicon content and adding 1.25 per cent nickel this hardness was increased to about 200 and the machining time on the bushings decreased, due to elimination of hard spots. Wear tests showed about 0.004 inches wear on the nickel iron bushings after the equivalent of 40,000 miles service as compared to 0.010 inches wear in the plain iron bushings.

COMPOSITION OF TRACTOR CYLINDER BUSHINGS

	Alloy Iron Per Cent	Plain Iron Per Cent
Nickel	1.25	
Silicon	1.90	2.25
Carbon	3.10–3.30	3.10–3.30

Improving Machinability Without Opening up the Structure or Impairing Hardness and Strength—The machinability of hard castings can be improved merely by increasing the silicon or carbon content of the

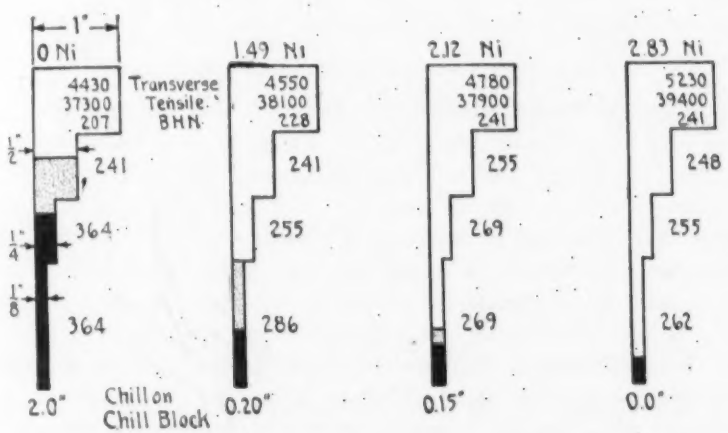


Fig. 4—Hardening Gray Iron by Adding Nickel. The Step Bar Test.

iron used, but at the expense of good structure and mechanical properties. When the latter must be preserved, nickel may be used to secure improved machinability. In this case from 0.50 to 1.0 per cent of nickel may be added to the plain iron used, and if desired the silicon content of the latter may at the same time be slightly reduced by 0.10 to 0.20 per cent.

3. Tractor Piston Rings—These rings (¼ to ½ inch in section) are cast as a bushing and machined from it, and must have a fine close grain. This requires about 2.10 per cent silicon in order that the center rings of the cast bushing will have the desired structure, but at this silicon content the edges are hard to machine. With a higher silicon content the edges are softer, but the center rings are too open in grain. The addition of 0.50 per cent nickel to the harder iron gave castings of satisfactory structure throughout, and on which the tool cost of machining was about 1/6 of that on the plain 2.0 per cent silicon iron.

COMPOSITION OF TRACTOR PISTON RINGS

	Alloy Iron Per Cent	Plain Iron Per Cent
Nickel	0.50	
Silicon	2.10	2.10
Carbon	3.35	3.35

4. Automobile Pistons—Pistons in plain iron tend to chill at the bottom of the skirt and have to be annealed at 1450 degrees Fahr. in order to machine. This softens them rather too much and they are readily distorted after annealing. Made in higher silicon iron the cross section may show internal shrinkage. By lowering the silicon content slightly and adding about 0.90 per cent nickel the piston castings were completely gray and machinable without annealing and required only a strain relief anneal at 900 to 1000 degrees Fahr.

COMPOSITION OF AUTOMOBILE PISTONS

	Alloy Iron Per Cent	Plain Iron Per Cent
Nickel	0.80–1.00	
Silicon	2.40–2.55	2.60–2.75
Carbon	3.40	3.40

Production of Sound, Machinable Castings and Elimination of Porosity—One of the commonest foundry problems is the production of sound pressure castings in gray iron. It is recognized that soundness in structure is best secured by the use of iron of close and fine grained texture, and conversely, that porosity in castings generally increases as the silicon content increases and the grain beomes more open. High silicon iron (over 2.25 per cent) is particularly prone to exhibit internal shrinkage and porosity. In the production of castings which are to withstand pressure, therefore, low silicon iron is generally desired, but with such iron difficulty is frequently encountered in machining.

The addition of nickel permits the use of low silicon iron for castings even of moderate section, and removes the difficulty of machining them when produced in plain, low silicon iron.

The method of using nickel in connection with any particular pressure casting problem depends on whether the difficulty encountered is in machining, or in rejections due to unsoundness. If the castings are sound in the plain iron used, but difficultly machinable, then nickel should merely be added to reduce chill and restore machinability, and usually in amounts from 0.50 to 1.50 per cent.

If the castings are unsound on test and foundry rejections are high, the silicon content of the iron should be reduced by 0.25 to 0.50 per cent and from 0.50 to 1.50 per cent nickel added to compensate (in point of chill) for the silicon reduction.

5. Ammonia Compressor Heads—These castings, about 1½ to 4½ inches in section, made in plain iron were showing about 7½ per cent rejection in foundry and machine shop, due to unsoundness. By lowering the silicon content and adding about 0.75 per cent nickel, the rejections were reduced to about 1 per cent.

Composition and Mixture of Compressor Heads

	Alloy Iron Per Cent	Plain Iron Per Cent
Nickel	0.72	....
Silicon	1.04	1.35
Carbon	3.08	3.14
Sulphur	0.085	0.12
Phosphorus	0.30	0.19
Mixture		
Steel scrap	250 lbs.	200 lbs.
Cast Iron scrap	500 lbs.	350 lbs.
No. 2 foundry pig	...	300 lbs.
Malleable Bessemer pig	200 lbs.	150 lbs.

6. *Molding Machine Cylinders*—These castings of sections ranging from ½ to 1½ inches have to pass a 100 pound air pressure test. When made with enough silicon (1.75 to 2.00 per cent) to render them readily machinable, they were often locally porous in structure and gave from 10 to 12 per cent rejections on test. By lowering the silicon to 1.25 to 1.50 per cent and adding a compensating amount of nickel, the rejections were reduced practically to zero and the machinability was improved slightly.

Composition and Mixture of Cylinders

	Alloy Iron Per Cent	Plain Iron Per Cent
Nickel	1.00	....
Silicon	1.25	1.75
Carbon	3.40	3.40
Phosphorus	0.29	0.43
Sulphur	0.35	0.125
Mixture		
Steel scrap	300 lbs.	250 lbs.
Cast Iron scrap	500 lbs.	500 lbs.
No. 2 x pig	...	250 lbs.
No. 2 Malleable Bessemer	200 lbs.	...

Securing Pearlitic Structure Free from Carbides—Full pearlitic structure in castings is generally desirable and is sought because of its better mechanical properties, particularly its better wear resistance as compared to iron partially ferritic in structure. Whether or not a casting will have pearlitic structure depends partly upon its section and cooling rate. Light sections show, in general, higher combined carbon than heavier sections cast of the same iron. Pearlitic structure depends principally upon composition and upon the silicon content of the iron, and is obtained with iron of low silicon content (0.75 to 2.0 per cent, depending upon section).

Castings which require such a low silicon content for full pearlitic structure as to render them difficultly machinable in plain iron, may be rendered readily machinable by the addition of 1.0 to 2.0 per cent nickel to iron of the required low silicon content. In extreme cases where iron of low chilling tendency is required for machining, but pearlitic structure is desired, advantage may be taken of the tendency of nickel in higher concentrations (1½ to 3½ per cent) to build up combined carbon even in high silicon irons.

7. *Automobile Cylinder Block*—Because of the presence of copper fins cast integrally in one design of cylinder block, it is necessary to use iron

of very low chilling power, as otherwise the metal adjacent to these fins is chilled and unmachinable. Such an iron, with about 2.50 per cent silicon, gives a structure in the cylinder bore which is about half pearlitic and half ferritic and consequently of inferior wear resistance.

By adding about 2.50 per cent nickel the structure is changed to practically full pearlitic, the Brinell hardness in the bore is increased by about 30 to 40 points and the structure of the metal chilled by the metallic inserts is rendered machinable.

Composition of Automobile Cylinder Blocks

	Alloy Iron Per Cent	Plain Iron Per Cent
Nickel	2.25–2.50	.....
Silicon	2.40–2.55	2.40–2.55
Carbon	3.40–3.50	3.40–3.50
Manganese	0.60–0.70	0.60–0.70
Sulphur	0.10	0.10
Phosphorus	0.50	0.50

Often the absence of excess carbides in the structure is just as important as the absence of free ferrite, both from the standpoint of machining and that of wear. (This statement obviously should not be construed as applying to white iron, the good wear resistance of which depends primarily on the presence of these carbides).

8. Clutch Plates—Plain Iron clutch plates (sections 9/16 to ⅝ inch) of low silicon iron gave inferior wear service (2 to 3 weeks) in spite of their high Brinell hardness (196). Investigation showed the structure to be pearlitic but with considerable free carbide which presumably acted as an abrasive and accelerated the ease of wear. About 1.50 per cent nickel was added to the iron, which reduced its chilling tendency and eliminated the excess carbides while still retaining the full pearlitic structure at about 203 Brinell. Clutch plates of this iron give over five months service, and the improved performance is attributed to the absence of the abrading particles of free carbide.

	Alloy Iron Per Cent	Plain Iron Per Cent
Nickel	1.25–1.75	...
Silicon	1.50	1.50
Carbon	3.20	3.20

Uniform Structure in Sections of Different Thickness—Castings having both heavy and light sections are difficult to produce with good structure in the heavy section and with machinable light section. Nickel added to a composition of iron low enough in silicon and carbon to give good structure in the heavy section will render the light section machinable.

9. Pressing Machine Steam Chest—This casting varied in section from ¼ inch on the edge to 1½ inches at the bosses. In 2.25 per cent silicon iron the edges were machinable but the bosses open and porous. In 1.50 per cent silicon iron the bosses were sound in structure, but the edges were unmachinable. Addition of 1.0 per cent nickel to the lower silicon iron gave a satisfactory casting in both respects.

Composition and Mixture

	Alloy Iron Per Cent	Plain Iron Per Cent
Nickel	1.00	....
Silicon	1.40	2.25
Carbon	3.30	3.40
Phosphorus	0.20	0.60
Sulphur	0.10	0.08
Manganese	0.65	0.45

Mixture

	Per Cent	Per Cent.
Steel in mixture	30	..
Cast Iron scrap	40	40
Pig Iron	..	60
Malleable Bessemer Pig	30	..

Refining the Grain of Machinable Castings—Fine grain structure is often required in iron castings, and for a variety of reasons, principally for good machine finish, free from flaws, and for satisfactory polish finish. Difficulty of machining generally limits the possibility of using low silicon iron in securing fine grained castings and in such cases nickel may be used to restore machinability.

10. Printing Press Rolls—Printing Press rolls (11 feet by 30 inches by 1½ inch wall) in plain 1.80 per cent silicon iron were machinable, but the center portions were open and porous in structure. In lower silicon plain iron the ends were hard to machine. A uniform and machinable structure was secured in a 1.40 per cent silicon iron with nickel added to secure machinability.

Composition of Printing Press Rolls

	Alloy Iron Per Cent	Plain Iron Per Cent
Nickel	2.00	....
Silicon	1.40	1.80
Carbon	3.30	3.30

Increasing Strength in Machinable Castings—The simple addition of nickel to iron is not recommended when increased strength is desired. For this purpose advantage is taken of the chill reducing effect of the nickel; the silicon, or even better, the total carbon content of the iron is lowered by using greater percentages of steel scrap, and nickel is added to restore machinability.

By lowering the total carbon to about 3.10 to 3.20 per cent and with appropriate silicon and nickel contents, tensile strengths (on arbitration bars) in the neighborhood of 40,000 pounds per square inch may be obtained in iron of sufficiently low chilling tendency to give machinable castings even of moderate section.

11. Locomotive Cylinders—By using about 15 per cent more steel in the charge and adding 0.50 to 0.75 per cent nickel it was possible to secure tensile strengths in iron for locomotive cylinders of 40,000 pounds per square inch as compared to 35,000 pounds per square inch for plain iron. At the same time the machinability of the cylinders (of sections 1½ to 2½ inches) was improved.

Composition of Locomotive Cylinders

	Alloy Iron Per Cent	Plain Iron Per Cent
Nickel	0.50–0.75	...
Silicon	1.00	1.25
Carbon	3.15	3.25

12. Superheater Castings—In the same manner a stronger iron was secured for superheater body castings (section ¾ to 2 inches) by increasing the steel in the charge and adding 1.00 per cent nickel. The plain iron showed 4,500 pounds transverse and 35,000 pounds per square inch tensile strength as compared to 5,000 and 40,000 pounds for the alloy iron. At the same time, the alloy iron was found to be somewhat more readily machined owing to the absence in its structure of hard spots and free carbide.

Composition and Mixture of Superheater Castings

	Alloy Iron Per Cent	Plain Iron Per Cent
Nickel	1.00	....
Silicon	1.00	1.20
Carbon	3.30	3.40
Phosphorus	0.16	0.19
Sulphur	0.08	0.08

Mixture

	Per Cent	Per Cent
Steel in mixture	43	35
Foundry scrap	35	35
Charcoal	..	30
Malleable Bessemer pig	22	..

Foundry Production of Nickel Iron—The production of nickel iron in the iron foundry follows usual practice. Nickel iron has the same melting point, the same contraction, and other casting characteristics as plain gray iron. Therefore the usual melting, molding, and casting practice needs no modification, except for the actual addition of nickel (and other changes in iron composition which may be made simultaneously; viz., lowered silicon content, increased per cent of steel in charge, etc.).

Nickel may be added either in the cupola (air or electric furnace) or in the ladle, and is available in several forms for this purpose as follows:

1. Pure nickel (99 per cent) either in 5 pound pigs or in shot (melting point 2640 degrees Fahr.).

2. Low melting point nickel (''F'' nickel)—5 pound pigs or in shot (melting point 2300 degrees Fahr.) containing 92 per cent nickel and 6.0 per cent silicon. Both forms are successfully used in pig form in the cupola (or air or electric furnace) and in shot form in the ladle. The low melting point nickel is more widely used, however, and dissolves more readily in ladle iron, particularly when using small ladles or cold iron.

Ladle iron should be at a temperature of not less than 2500 degrees Fahr. when nickel is added in order to insure uniform solution. When pig nickel is charged into the cupola it should be placed as near the center of the charge as possible. Obviously the usual precautions should be taken to separate nickel bearing charges from non nickel bearing ones.

Nickel shot may be added to ladle iron by a funnel attached to the cupola, which delivers the weighed amount of nickel directly to the metal issuing from the tap hole or the shot is dropped from a trough or tube directly into the stream of iron being poured into the ladle.

It is essential to add the shot in such a manner that it will be dragged under the surface of the molten iron and stirred into the body of iron and to avoid adding it in such a way that it merely floats on the surface or lies on the bottom of the ladle, sinters together and does not dissolve.

TENTATIVE RECOMMENDED PRACTICE FOR HEAT TREATMENT OF CHISELS*

General—Chisels must meet severe fatigue stresses and retain good cutting edges. Various steels as well as heat treatments are employed to meet these conditions. Only the principal steels and simpler treatments are given here.

Approximate Chemical Composition

Steel No.	Type	C	W	Cr	V
1	Carbon	.70–.95	....	....	...
2	C-V	.70–.95	....	....	.20
3	Cr-V	.45–.60	....	.80	.20
4	Cr	.55–.70	....	.50	...
5	W-Cr-V	.45–.55	2.00	1.25	.25

Heat Treatments for Chisels

Steel Nos. as Listed in Table I	Forging Temperatures[1] Deg. Fahr.	Hardening Temperatures Deg. Fahr.	Quenching Media	Tempering Temperatures Deg. Fahr.
1	1600–1400	1400–1475	Water	350–550
2	1600–1400	1400–1475	Water	350–550
3	1650–1400	1450–1500	Water	350–550
4	1650–1400	1450–1500	Water	350–550
5	1700–1400 1700–1400	1650–1700 1550–1600	Oil Water	450–550 450–550

Note—With pneumatic chisels various hardening methods are followed to put the shank in proper condition. Sometimes the chisel is treated as above and sometimes the shank is given a preliminary quench in oil from the temperatures given in the table, after which the cutting edge is hardened as given in the table, and the whole chisel tempered at the temperatures given.

---

[1]The initial forging temperature depends upon the amount of work to be done upon the steel and in no case should the finishing temperature be below 1400 degrees Fahr.

---

*This is a—Tentative Recommended Practice—and will remain tentative for at least one year, until adopted by the Board of Directors and the Recommended Practice Committee of the A. S. S. T. This practice is not intended for a specification and should not be interpreted as such.

The membership of the Sub-Committee on the Heat Treatment of Tool Steels is as follows: J. P. Gill, Chairman, J. A. Succop, W. H. Phillips, A. D. Beeken, Jr., Joseph Taylor, C. M. Johnson, Du Ray Smith, A. M. Cox, and Frank Garratt.

Criticisms of this Tentative Practice are solicited and should be directed to J. E. Donnellan, Secretary of the Committee, 7016 Euclid Avenue, Cleveland, Ohio.

CONSTITUTION OF NICKEL-MANGANESE ALLOYS

By N. B. Pilling* and T. E. Kihlgren*

The nickel-manganese series has received little attention from investigators even though the lower manganese content alloys are of some technologic importance. The early diagram of Zhemezhuzhny, Urasov and Rykovskov,[3] which represents this as a continuous series of solid solutions with a minimum in the liquidus has received general acceptance, although

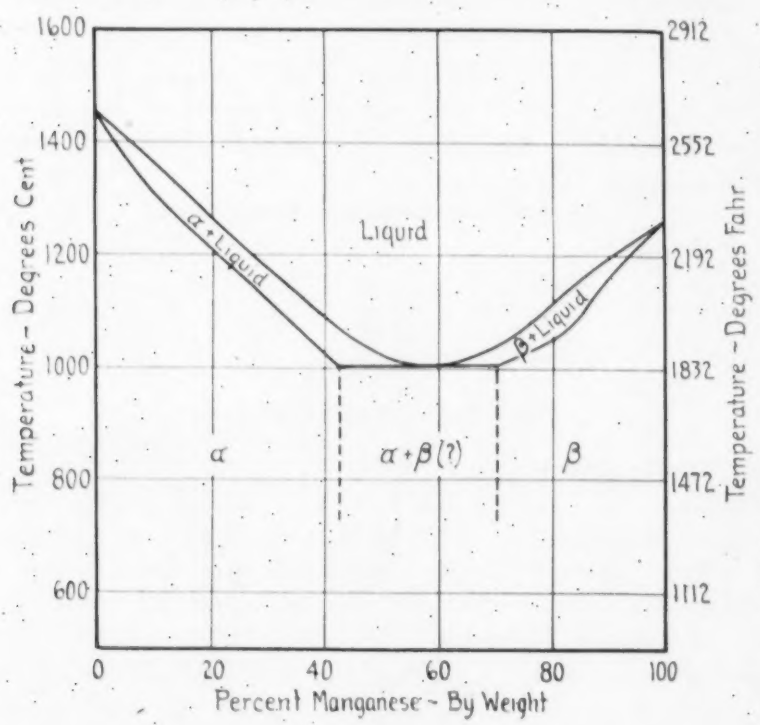


Fig. 1—Nickel-Manganese Equilibrium Diagram.

there seem to be some reasons for believing that this diagram is incorrect. Nickel crystallizes with a face-centered cubic lattice and the high temperature β modification of manganese in a more complex cubic lattice.[1] Evidently isomorphous mixtures are not to be expected. Dourdine,[2] confirming the general aspects of the liquidus curve with a purer set of alloys, found in fact microscopic evidence of a gap in solid miscibility in the range 38-71 per cent manganese, although this evidence was not incorporated in his diagram.

The present diagram shown in Fig. 1 is based largely on Dourdine's data and has been drawn to represent a pair of solid solutions with limited miscibility.

---

*The International Nickel Co., Inc., Bayonne, N. J.

Prepared for the Nonferrous Data Sheet Committee of the Institute of Metals Division of the A. I. M. E. and the Recommended Practice Committee of the A. S. S. T.

No attempt has been made to indicate other than suggestively the equilibrium within the heterogeneous range. Gayler[1] has recently found thermal evidence of four transformations in the solid state in manganese of exceptionally high purity at 1191, 1024, 742, and 682 degrees Cent. The third transformation is possibly concerned with a structural change. Both investigators of the alloys[2,3] have found thermal disturbances occurring between 500 and 1000 degrees Cent. in the cooling curves of alloys within the heterogeneous range, Dourdine reporting no less than six series of transformations of uncertain significance. Their relation to the manganese transformations is not unlikely and it is suggested that the alloys containing more than 40 per cent manganese may be of considerable complexity.

References

1. M. L. V. Gayler, *Journal*, Iron and Steel Institute, Vol. 115, 1927, p. 393.
2. A. D. Dourdine, *Revue de Metallurgie*, Vol. 12, 1915, p. 125.
3. S. Zhemczhuzhny, G. Urasov and A. Rykvoskov, *Zeitschrift für Anorganische Chemie*, Vol. 57, 1908, p. 253.

Reviews of Recent Patents

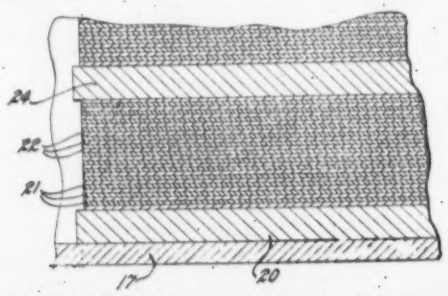
By NELSON LITTELL, Patent Attorney
475 Fifth Ave., New York City—Member of A. S. S. T.

1,706,438, March 26, 1929, Method of Annealing Silicon-Steel Sheets, William E. Caugherty, of Natrona, and Carl J. Stroble, of Tarentum, Pennsylvania, Assignors to Allegheny Steel Company, a corporation of Pennsylvania.

This patent describes the method of annealing silicon steel sheets of light gage so as to secure the best magnetic and electrical properties with freedom from brittleness, etc. The inventors have discovered that to secure this result the sheets should be rapidly heated to a temperature for best electrical conductivity and maintained at that temperature for the desired period of time. Rapid heating of the sheets is secured by charging one, two or three sheets at a time into a furnace and waiting a sufficient period of time before additional sheets are charged. The additional sheets are charged in lots of one, two or three sheets at a time until the stack is built up to the desired height and capacity of the furnace and is then left in the furnace for approximately three hours time, after which the furnace is allowed to cool and the stack of sheets removed.

1,706,725, March 26, 1929, Method of Heat Treating Articles, Austin Le Boutillier, of Western Springs, Illinois, Assignor to Western Electric Company, Incorporated, of New York, N. Y., a corporation of New York.

This patent describes a method of heat treating "permalloy" sheets so as to prevent adherence of the sheets to one another. The method comprises packing the sheets 21 of "permalloy" with interleaved sheets 22 of silicon steel having an oxide coating on the surface thereof which will not adhere to the "permalloy" sheets at high temperature. The sheets are packed between relatively heavy flat steel plates 20 and 24 and are weighted down by additional plates, so as to prevent distortion or buckling during the heat treatment.



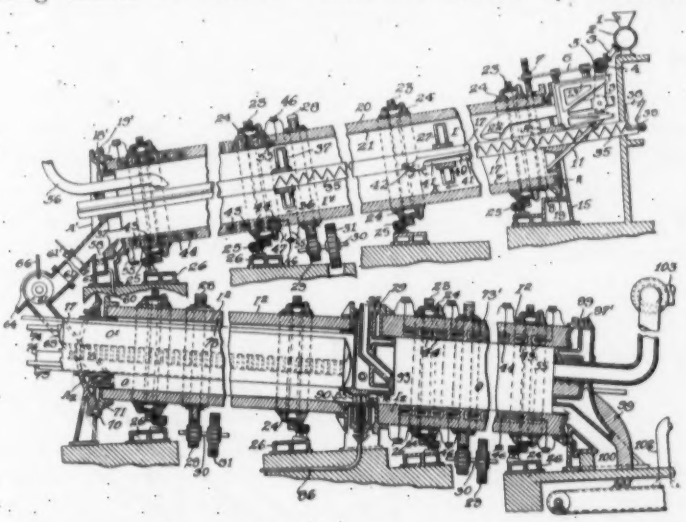
1,706,858, March 26, 1929, Process of Treating Molds for Producing Soft Castings, Hubert A. Myers, of Toledo, Ohio, Assignor to The Metal Castings Holding Company, of Cleveland, Ohio, a corporation of Delaware.

This patent describes a method of treating permanent molds for the

production of machinable ferrous castings therein so as to prevent pitting of the mold surface and prevent chilling of the surface of the iron alloy poured into the molds. The process comprises coating the matrix of the mold with a material containing potassium cyanide and heating the mold and the coating to cause decomposition of the potassium cyanide and penetration of the carbon into the face of the mold. It is preferable to also use a thin coating of heat insulating material, such as plumbago or commercial magnesium oxide.

1,708,526, April 9, 1929, Rotary Chemical Furnace, Luigi Tocco and Michele Landi of Paris, France, Assignors to Societe "Le Four Chimique Rotatif" (Societe Anonyme Francaise,) of Paris, France.

This patent describes a rotary chemical furnace consisting of the inclined rotaing shafts 20 and I² which are heated by electrical heating



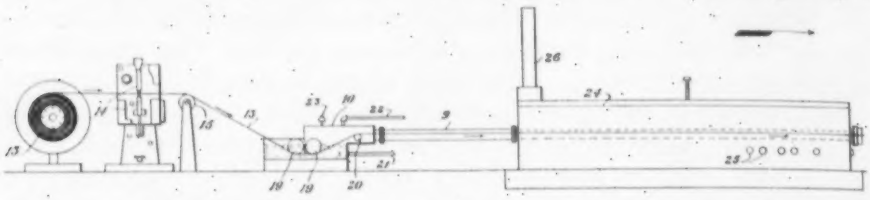
elements 43 in channels 44. The material undergoing treatment is fed into the furnace from a hopper 1 and passes by gravity through the furnace through the communicating chutes 61 into the lower furnace. Carbon or other reducing material may be also fed into the furnace through the hosper 38 and distributed at the desired points therein by the screw conveyor 35. Various driving and steam generating means are described in the specification.

1,714,038, May 21, 1929, Process of Treating Silicon Steel, Alphonsus F. Murphy, of Zanesville, and William Jones, of Middletown, Ohio, Assignors to The American Rolling Mill Company, of Middletown, Ohio, a corporation of Ohio.

This patent describes a process of producing smooth finished, scale-free high silicon steel sheets for electrical purposes. The process comprises passing the manufactured sheets through a pickling treatment of sulphuric acid, controlling the solution to prevent pitting, neutralizing the sheets by a suitable base, imposing a mechanical separator on the sheets and annealing the sheets and finally giving a smooth surface to the sheets by cold rolling, followed by a second annealing.

1,714,040, May 21, 1929, Continuous Annealing Process, Harry M. Naugle and Arthur J. Townsend, of Canton, Ohio, Assignors, by Mesne Assignments, to The Columbia Steel Company, of Pittsburgh, Pennsylvania, a corporation of Pennsylvania.

This patent describes a continuous annealing process in which the strip 13, undergoing annealing, is passed through a water seal under the rollers 19-19′ at each end of the apparatus before entrance into the annealing tube 9. The annealing tube 9 is heated for a portion of its length by the



furnace 24 and the heating is carried out in the presence of a non-oxidizing gas. The strip 13 is cooled to a temperature of less than 100 degrees Cent. while still in the non-oxidizing gas and passed out of the annealing tube 9 through another water seal. The water seal used is of advantage over a sand seal, which abrades and defaces the surface of the metal, and also over an oil seal which coats the strip 13 with an adhering film which is objectionable in the annealing chamber.

1,713,766, May 21, 1929, Alloy, George G. Marshall and Harold S. Booth, of Cleveland, Ohio.

This patent describes a means of reducing the high specific gravity of chromium alloys by alloying therewith beryllium which, in addition to lowering the specific gravity of the alloy, forms with the chromium an adherent skin oxide which protects the underlying metal against deterioration, due to a further oxidation. The preferred composition of the alloy consists of chromium 6 per cent, iron 90 per cent and beryllium 4 per cent.

1,714,729, May 28, 1929, Metal, Frederick J. Read, of San Francisco, California, Assignor, by Direct and Mesne Assignments, of One-Half to Calaveras Iron and Steel Company, a corporation of Nevada, and One-Half to Sidney H. Demarest, of San Francisco, California.

The object of the present invention is to provide a new metal which is particularly adaptable for use as bearings. The alloy includes a combination of cerium and aluminum with copper. The metal radiates heat rapidly from the journal bearing so that relatively close clearances can be used and is capable of running for long periods with a minimum amount of lubrication.

1,715,867, June 4, 1929, Vanadium Alloy, Byramji D. Saklatwalla, of Crafton, Pennsylvania, Assignor to Vanadium Corporation of America, of Bridgeville, Pennsylvania, a corporation of Delaware.

This patent describes a ferrovanadium alloy containing from 85 to 95 per cent of vanadium, which is more advantageous in the preparation of vanadium steels than a lower vanadium content in the alloys presently used. The alloy is produced in an electric furnace by the use of a reducing agent, such as silicon.

THE ENGINEERING INDEX

Registered United States, Great Britain and Canada.

In the preparation of the Engineering Index by the staff of the American Society of Mechanical Engineers some 1700 domestic and foreign technical publications received by the Engineering Societies Library (New York) are regularly searched for articles giving the results of the world's most recent engineering and scientific research, thought, and experience. From this wealth of material the A. S. S. T. is supplied with this selective index to those articles which deal particularly with steel treating and related subjects.

AIRPLANE FITTINGS

DESIGN. Fundamentals of Airplane Fitting Design, R. C. Gazeley, *Aviation*, vol. 25, no. 11, Mar. 16, 1929, (Aeronautical Eng. Soc.) pp. 34-39, 8 figs.

Fundamental principles of fitting design are discussed; overcoming eccentric moment which tends to rotate fitting; bolted joints, unless reamed to perfect fit and provided with ample bearing area, will inevitably pound out; 20 per cent margin of safety required; selection of most efficient material; heat treatment; effect of repeated bending loads; riveting and welding.

AIRPLANE PROPELLERS

MANUFACTURE. The Duralumin Propeller, G. Svehla. *Popular Aviation and Aeronautics*, vol. 4, no. 3, Mar. 1929, pp. 48, 50, 52, 54 and 97, 10 figs.

Advantages of duralumin propellers and procedure in manufacture blades for Army and Navy are outlined; practically all blades forged or rolled; forging of blade for two-piece propeller; aging to give added hardness to metal.

AIRPLANE MANUFACTURE

METAL CONSTRUCTION. The Metal Version of the Westland "Widgeon" Light 'Plane. *Flight, (Lond.)*, vol. 21, no. 9, Feb. 28, 1929, p. 168, 12 figs.

Constructional features of metal fuselage of Westland Widgeon are illustrated in 12 line cuts, with brief descriptions.

WELDING BLOWPIPES. Aircraft Blowpipe. *Acetylene Jl.*, vol. 30, no. 8, Feb. 1929, pp. 340 and 342, 1 fig. See also *Sheet Metal Worker*, vol. 20, no. 3, Feb. 1929, p. 88.

Brief description of Oxweld Type W-15 aircraft welding blowpipe produced by Oxweld Acetylene Co., New York, and especially designed for airplane fuselage welding.

Case Hardening of Steel (Nitridation of Steel), N. A. Minkevich and I. I. Kontrovich. *Vestnik Metallopromyschlennosti (Moscow)*, vol. 8, nos. 7 and 8, 9 and 10, July and Aug., Sept. and Oct., 1928, pp. 5-25 and 69-84, 41 figs.

Review of literature and theories on steel-nitridation process: report on authors' experimental study made in Russia's Aviation Trust laboratories for purpose of developing most suitable alloy steels for construction of measuring apparatus and of machine parts subject to intense wear and frequent reversal of stresses. (In Russian.)

CORROSION. Corrosion and Rusting of Cast Iron With and Without Alloys (Ueber die Korrosion und das Rosten von unlegiertem und legiertem Gusseisen), P. Koetzschke and E. Piwowarsky. *Stahl und Eisen (Duesseldorf)*, vol. 48, no. 49, Dec. 6, 1928, p. 1716.

Brief abstract of paper previously indexed from Archiv fuer das Eisenhuettenwesen, Nov. 1928.

ALLOYS

HEAT RESISTING. Heat-Resistant Alloys. *Am. Mach.*, vol. 69, no. 21, Nov. 22, 1928, p. 817.

Calite, Chromel, Fahralloy, Hybnickel, Q-alloys, Rezistal, and Silcrome are described.

WELDING. Aluminum Welding, F. Grove-Palmer. *Metal Industry (Lond.)*, vol. 33, no. 22 and 23, Nov. 30 and Dec. 7, 1928, pp. 509-510 and 533-534 and 546, 3 figs.

There is different technique for welding sheet and castings, for castings are more brittle, and with increased thickness there is greater liability of cracking due to expansion and contraction during working; annealing zones are weakest parts of job; care is required for duralumin; far better results obtained from cylinder gas, though more expensive than generator gas; flux should consist of chemical salts that will melt earlier than duralumin and will dissolve oxide.

ALUMINUM

MACHINING. Tools and Methods for Machining Aluminum, R. L. Templin. *Machy. (N. Y.)*, vol. 36, no. 7, Mar. 1929, pp. 557-559, 4 figs.

Detailed directions for making tools for turning, facing, planing and milling aluminum; cutting lubricants or coolants for machining aluminum; keen cutting edges important on cutting tools.

Those members who are making a practice of clipping items for filing in their own filing system may obtain extra copies of the Engineering Index pages gratis by addressing their request to the society headquarters, whereby their names will be placed on a mailing list to receive extra copies regularly.

Photostatic copies (white printing on a black background) of any of the articles listed may be secured through the A. S. S. T. The price of each print, up to 11 by 14 inches in size, is 25 cents. Remittances should accompany orders.

PROPERTIES. Tests of Effects of Melting, Rolling, and Heat Treatment on Properties of Aluminum (Untersuchungen ueber den Einfluss des Umschmelzens sowie des Walz und Gluehprozesses auf Aluminum), E. Maass and W. Wiederholt. *Korrosion und Metallschutz (Berlin)*, vol. 4, no. 12, Dec. 1928, pp. 272-277.

Report from division of metal chemistry and metal production of German Government Institute of Engineering Chemistry; effect of rolling and heat treatment on chemical properties, specific gravity, hardness, cupping quality, resistance to chemical corrosion, etc.

ALUMINUM BRONZE CASTING

Aluminum Bronze Cast In Permanent Molds, W. R. Williams. *Am. Mach.*, vol. 70, no. 13, Mar. 28, 1929, pp. 503-504, 2 figs.

Methods employed by Westinghouse Electric and Mfg. Co. for casting parts weighing up to 7 lb. and of wide variety of shapes, using alloy with small percentage of iron; sectional molds not desirable because of spreading at joints.

ALUMINUM CASTINGS. Advantages of Aluminum as a Foundry Material, G. Mortimer. *Can. Foundryman (Toronto)*, vol. 20, no. 3, Mar. 1929, pp. 21-23.

Advantages of aluminum castings and their properties in regard to weight reduction, machining, finish, and cheap cost of casting; aluminum alloys; properties of 2L-5; 3L-11 for high temperature; L-8 for water pistons; choice of alloys.

ALUMINUM METALLOGRAPHY

The Slip-Bands Produced When Crystals of Aluminum are Stretched-Extension at High Temperatures, K. Yamaguchi and S. Togino. *Inst. Phys. and Chem. Research—Sci. Papers (Tokyo)*, vol. 9, no. 172, Jan. 10, 1929, pp. 277-292, 14 figs.

Description of experiments made at high temperatures in order to confirm quantitive relation found to exist between resistance to slipping of crystals and slip bands; orientation of crystal axes; surface of stretched specimen reveals parallel strips closely resembling ordinary slip-bands; hardness after extension; extension in liquid air. (In English.)

ALUMINUM SILICON ALLOYS

Aluminum-Silicon Alloys: Their Properties and Some Application, R. B. Deeley, *Foundry Trade, Jl. (Lond.)*, vol. 40, no. 650, Jan. 31, 1929, pp. 87 and (discussion), 87-90, 15 figs.

Application of aluminum-silicon alloys are illustrated, with brief descriptive notes in text.

AUTOMOBILE PLANTS

FOUNDRIES. Automobile Foundry Pours Aluminum Castings on a Conveyor, P. Dwyer. *Foundry*, vol. 57, no. 6, Mar. 15, 1929, pp. 222-227 and 230, 12 figs.

Description of new aluminum foundry of Packard Motor Car Co., Detroit and methods employed there; two conveyor units on which all molds are assembled and poured; complete sand handling outfit prepares sand and keeps it in circulation; extensive monorail system for handling metal and molds; two large reservoir furnaces and six tilting pot furnaces.

MATERIALS HANDLING. Cost of Cylinder Block Machining Reduced by Graham-Paige Through Lessening of Manual Handling, A. F. Denham. *Automotive Industries*, vol. 60, no. 11, Mar. 16, 1929, pp. 434-438, 6 figs.

Description of conveyors, hoists, and straddling machines used in machining line for engine blocks of smaller cars, which is designed to handle 256 units per day; location of all machining operations from part to which they are most closely related.

BEARING METALS, BRONZE

Testing Bearing Bronzes. *Elec. Traction*, vol. 25, no. 2, Feb. 1929, pp. 74-75, 2 figs.

Study of heating chill-cast and sand-cast conditions yields interesting facts regarding wear and mechanical properties; charts showing representative specifications for journal bearings.

PROPERTIES. Mechanical Properties of Bearing Bronzes. *Mech. World (Manchester)*, vol. 85, no. 2196, Feb. 1, 1929, p. 104.

Discussion of report of United States Bureau of Standards upon wearing and mechanical properties of two groups of copper-tin-lead alloys in both chill-cast and sand-cast condition; result of tests under rolling and sliding friction without lubricants at atmospheric and elevated temperatures.

BERYLLIUM ALLOYS

Beryllium, a New Light Metal (Beryllium, ein neues Leichtmetall) Freitag. *Motor (Berlin)*, vol. 17, no. 1, Jan. 1929, pp. 66-67.

Beryllium has only recently been considered for commercial uses; impracticable to use it as pure metal because of its price; it is lighter than aluminum; iron with 2 per cent beryllium will attain 3 times its normal hardness and with appropriate heat treatment, 6 times normal hardness; iron with 4 per cent beryllium cannot be cold-rolled; copper with 6 per cent beryllium has hardness of hardened steel; these alloys are practically immune to corrosion and atmospheric effects.

BRONZE FOUNDRY PRACTICE

Utilization in Bronze Foundries of Silica-Sand Cores (Essais pour l'utilisation en fonderie de bronze du noyautage en sable siliceux agglomere), A. Miroux. *Fonderie Moderne (Paris)*, vol. 23, Feb. 25, 1929, pp. 72-74, 4 figs.

Results of tests carried out by author to investigate defects which frequently occur in connection with use of silica-sand cores in bronze foundries; influence of diameter on cores; influence of thickness and composition of metal; effect of drying.

CADMIUM PLATING

Cadmium Coating, G. De Lattre. *Metallurgist (Supp. to Engineer, Lond.)*, Mar. 1929, pp. 41-42.

Coatings of pure cadmium are dealt with first; metal offers advantage of lower melting point (about 320 deg. cent. as compared with 425 deg. for zinc) which means easier working, less wear and tear on pots and lower cost in fuel; protective effect of cadmium and its own behavior in regard to corrosion; use of alloys of cadmium in place of pure

metal. Abstract translated from Revue de Métallurgie, Nov. 1928.

Electro-Deposition of Cadmium for Rust Prevention, S. Wernick. *Metallurgist (Supp. to Engineer, Lond.)*, Mar. 1929, pp. 36-38.

Notes on composition of solutions; electrode-efficiency ratio; effects of variations in composition of solution; properties of deposit; composite deposits; mechanical properties. Review of paper read before Electroplaters and Depositors' Tech. Soc.

The Electrodeposition of Cadmium for Rust Prevention, S. Wernick. *Metal Industry (Lond.)*, vol. 34, no. 10, Mar. 8, 1929, pp. 245-248 and (discussion) no. 11, Mar. 15, pp. 277-280.

Wide divergence exists in composition of cadmium-plating solutions recommended by different authorities; electrode efficiency ratio; effects of variations in composition of solution by increasing cadmium content, temperature and current density; throwing power of solution; use of cadmium plating as under-coating to more pleasing finishes of less protective capacity. Read before Electroplaters' and Deposits' Tech. Soc.

CHROMIUM PLATING

Chromium Plating Steel Before Heat Treating, S. Tour. *Rochester Engr.*, vol. 7, no. 10, Apr. 1929, pp. 198-201, 13 figs.

Chromium plating gives promise of suitable protective layers that can be put on steel direct, to protect it from oxidation, carbonization or decarbonization during heat-treatment process.

How Chromium Plating Will Reduce Wear, W. Blum. *Traction Shop and Roadway*, vol. 2, no. 3, Mar. 1929, p. 75.

Brief abstract of paper presented before Am. Soc. Mech. Engrs. previously indexed from Machy. (Lond.), Jan. 3, 1929.

COPPER ALLOYS

Some Notes on Non-Ferrous Casting Alloys, W. Lambert. *Can. Foundryman (Toronto)*, vol. 20, no. 3, Mar. 1929, pp. 29-30.

Properties and casting practice for copper-aluminum bronzes and copper-tin group of alloys, true bronzes, including various types of red brass, gunmetal, and phosphor-bronze alloys, are described; Admiralty gunmetal; aluminum-bronze alloys susceptible to heat treatment; defective castings. From address read before Inst. of British Foundrymen.

WELDING. Welding Copper, Brass and Bronze, I. T. Hook. *Welding Eng.*, vol. 14, no. 3, Mar. 1929, pp. 45-48, 1 fig.

Welding characteristics and strength of alloys of copper, tin, and zinc in various proportions and procedure for successful welding by torch or metallic arc; any of copper-tin-zinc series of alloys are readily weldable when used as base metal; metal arc and carbon arc as well as oxyacetylene methods applicable to welding copper or copper-tin alloys; only indirect arc or oxyacetylene method applicable with zinc alloys. Paper presented before Welding Conference at Iowa State College.

CASTINGS. Tin-Copper Alloy Castings (Zur Kenntnis der gegossenen Zinn-Kupfer-Legierungen), W. Claus and H. Goeke, *Giesserei (Duesseldorf)*, vol. 16, nos. 4 and 5, Jan. 25 and Feb. 1, 1929, pp. 73-80 and 98-105 41 figs.

Jan. 25: Investigations of structure and mechanical and technical properties of standard, cast tin-copper alloys; influence of different molding materials and mold design on values of so-called reverse-ingot segregation of alloys; influence of heat variations. Feb. 1: Theory of reverse-ingot segregation.

Tin-Copper Alloy Castings (Zur Kenntnis der gegossenen Zinn-Kupfer-Legierungen), W. Claus and H. Goeke. *Giesserei (Duesseldorf)*, vol. 16, nos. 6 and 7, Feb. 8, and 15, 1929, pp. 125-132 and 153-160, 35 figs.

Feb. 8: Extent of ingot segregation is investigated; results of experiments to determine tensile strength and elongation; Brinell hardness. Feb. 15: Density testing. Bibliography.

STRUCTURE. Structure of Copper-Tin Alloys (Sur la structure des alliages cuivre-étain), W. Broniewski and B. Hachiewicz, *Revue de Métallurgie (Paris)*, vol. 26, no. 1, Jan. 1929, pp. 20-28, 13 figs.

Study of beta and gamma phases; tabular data on hardened alloys; curves of electric conductivity, thermoelectric properties, expansion, etc., are presented; photomicrographs.

DROP FORGING SHOPS

Prospects Bright in Drop Forgings, J. H. Williams. *Iron Age*, vol. 123, no. 1, Jan. 3, 1929, p. 51.

Drop-forging shops have met rising exactions of buyers but industry needs cooperative effort; expansion keeps wide gap of excess capacity; attention centered more than ever upon short cuts that lead to greater production; improvements in greater ruggedness, more generous use of steel casting, roller bearings, inserted guides, and increased ratio of weight between anvil and ram; aircraft demand of measurable importance.

ELECTRIC EQUIPMENT

DESIGN. Electrical Machinery and Apparatus, K. Sachs. *Brown Boveri Rev. (Baden, Switzerland)*, vol. 16, no. 1, Jan. 1929, pp. 3-34, 48 figs.

Synchronous, induction, and d.c. machines; industrial drives; transformers; mercury-arc power rectifiers; apparatus; electric furnaces; electric traction.

ELECTROPLATING—ALUMINUM

Note on the Testing of Electrodeposits on Aluminum, G. B. Brook, and G. H. Stott. *Inst. of Metals—Advance Paper (Lond.)*, no. 484, for mtg. Mar. 13-14, 1929, 9 pp., 9 figs.

Investigations carried out on samples of commercial electroplating on aluminum; most of these were finished in nickel; paper is confined to development of test, which indicate stability of deposit; it is rather from user's than from producer's standpoint that tests have been evolved.

IRON AND STEEL CORROSION

Report of Committee A-5 on Corrosion of Iron and Steel. *Am. Soc. Testing Matls.—Preprint*, no. 12, for mtg. June 24-28, 1929, 23 pp., 14 figs., partly on supp. plates.

Report on inspection of Annapolis tests, and on total immersion tests; report on field tests of metallic coatings.

Methods of Determination of Sulphur, R. P. Hudson. *Heat Treating and Forging*, vol. 15, no. 5, May, 1929, pp. 570-582.

Accurate and rapid methods for controlling sulphur in iron and steel; annealing samples often helpful; U. S. Bureau of Standards gravimetric method; oxidation, evolution, volumetric and iodine methods; preparation and standardization of iodine solution.

LEAD METALLURGY

Status of Lead Metallurgy (Etat actuel de la métallurgie du plomb), M. Fourment. *Revue de Métallurgie (Paris)*, vol. 26, no. 3, Mar. 1929, pp. 154-167, 1 fig.

Lead production has developed considerably in past few years; very few new deposits have been discovered; data on production in Europe, Asia, Africa, Australia, and America; technological aspects; treatment of mixed ores; refining of lead; hydrometallurgy; electrolytic treatment.

MACHINE TOOLS

CHROMIUM PLATING. Chromium Plating for Wear Resistance, W. Blum. *Machy. (N. Y.)*, vol. 35, no. 5, Jan. 1929, pp. 355-357.

Review of present-day applications of chromium plating to measuring devices and gages; hardness of chromium plate; adherence of chromium to object plated; success of chromium plating thickness of deposit on chromium-plated gages; changes in dimension produced by chromium plating on thread gages. (To be continued.) Abstract of paper read before Am. Soc. Mech. Engr.

CHROMIUM PLATING. Chromium Plating for Wear Resistance, W. Blum. *Machy. (N. Y.)*, vol. 35, no. 6, Feb. 1929, pp. 430-432.

Application of chromium plating to drawing, forming, stamping, molding, and blanking dies; value of chromium plate on tools for cutting metals; wear-resisting properties of chromium-plated machine parts subjected to severe wear; one of most successful applications of chromium plating for wear resistance is on dies used for molding plastic materials, such as rubber, bakelite, and clay products. (Continuation of serial.) Abstract of paper read before Am. Soc. Mech. Engrs.

MAGNESIUM ALLOYS

Ternary Alloys of Magnesium, Aluminum and Zinc Rich in Magnesium, A. M. Bochvar and O. E. Kestner. *Vestnik Metallopromishlennosti (Moscow)*, no. 11, Nov. 1928 pp. 5-10, 13 figs.

Results of experimental metallographic study, including check on Meissner's and Chadwick's studies of binary system MgZn. (In Russian.)

MAGNESIUM ALLOY CASTINGS

Production and Properties of Magnesium Alloy Castings, E. Player. *Metal Industry (Lond.)*, vol. 33, nos. 24 and 25, Dec. 14 and 21, 1928; pp. 568 and 570 and 591-592.

Dec. 14: Paper is confined mainly to manufacture, properties and applications of magnesium alloy castings; composition of various Elektron alloys. Dec. 21: Physical properties of Elektron alloys; all magnesium alloys possess excellent machining qualities; use in foundry practice, as possible flux, and in molding and coremaking. (To be concluded.) Abridged from paper read before Co-ordinated Societies.

MOLYBDENUM STEEL—PERMANENT MAGNETS

Investigations of Molybdenum Steels to Determine Their Adaptability as Permanent Magnets (Untersuchungen an Molybdaenstaehlen zur Pruefung ihrer Verwendbarkeit als Dauermagnete), A. F. Stogoff and W. S. Messkin. *Archiv fuer das Eisenhuettenwesen (Duesseldorf)*, vol. 2, no. 9, Mar., 1929, pp. 595-600, 16 figs.

Determination of magnetic properties of steels with 0.16 to 4.48 per cent molybdenum and 0.65 to 1.5 per cent carbon in relation to hardening temperature when hardening in water and oil; aging; comparison with chromium and tungsten steels.

NICKEL-COPPER ALLOYS

Study of Nickel-Copper Alloys (Etude des alliages nickel-cuivre), A. Krupkowski. *Revue de Metallurgie (Paris)*, vol. 26, no. 3, Mar., 1929, pp. 131-153, 28 figs.

Historical review of research; preparation of alloys; magnetic properties; electric resistance; assumptions based on results of measurements. (To be continued.)

NITRIDATION

Surface Hardening of Special Steels with Ammonia Gas Under Pressure, R. H. Hobrock. *Am. Soc. Steel Treating—Trans.*, vol. 15, no. 4, Apr., 1929, pp. 543-557 and (discussion) 557-568, 15 figs.

Samples of special steels have been nitrided with ammonia gas under pressure and nature of case so produced has been examined; results show that increase in pressure of ammonia gas results in decrease in surface hardness but greatly increases depth of case produced in given time; curves are presented correlating hardness, time of treatment, and depth of case; explanation of phenomena observed.

NONFERROUS ALLOYS

SEGREGATION. Inverse Segregation, M. Ballay. *Metallurgist (Supp. to Engineer, Lond.)*, Nov. 30, 1928, pp. 165-167.

Review of papers published in Aug. and Sept. issues of Revue de Métallurgie, in which question is studied both experimentally and theoretically; experiments were made on several series of alloys, among them lead and tin; comparatively slight changes of concentration found even in vertical-tube experiments, and especially slowness with which they are established, appear to support view that Ludwig-Soret effect does not play large part in production of inverse segregation.

NONFERROUS FOUNDRY PRACTICE

Practical Non-Ferrous Melting Problems Aired. *Iron Age*, vol. 123, no. 16, Apr. 18, 1929, pp. 1081-1082. See also *Iron Trade Rev.*, vol. 84, no. 16, Apr. 18, 1929, p. 1070.

Review of Non-Ferrous Session of the American Foundrymen's Association Convention; practical problems in brass foundry discussed by H. J. Roast; casting large bronze door; making white bronze grills; operation and cost of electric melting; melting cost per ton; cost data declared high though satisfactory.

OPEN-HEARTH FURNACES

DESIGN. The Economic Size of the Open

Hearth. F. A. King. *Min. and Met.*, vol. 9, no. 264, Dec. 1928, pp. 550-551.

Discussion as to how big furnace can be operated successfully; experience with 300-ton furnaces at Weirton, W. Va., indicates that economic limit has not yet been reached. Paper is presented at Open-Hearth Conference Pittsburgh, Pa.

REGENERATORS. Calculation of Regenerators Based on Heat-Transmission Coefficient (Die Berechnung der Waermespeicher auf Grund der Waermedurchgangszahl), K. Rummel. *Stahl und Eisen (Duesseldorf)*, vol. 48, no. 49, Dec. 6, 1928, pp. 1712-1715, 3 figs.

Attempt is made to calculate certain important values of regenerators according to formulas for recuperators; degree of temperature fluctuation during adjustment period; influence of brick thickness; general recommendations for design.

REGENERATORS. Influence of Dust Deposits on Efficiency of Checkerbrick Regenerators (Der Einfluss des Staubbelags auf den Wirkungsgrad von Gitter-Waermespeichern), A. Schack. *Archiv fuer das Eisenhuettenwesen (Duesseldorf)*, vol. 2, no. 5, Nov. 1928, pp. 287-291 and (discussion) 291-292, 2 figs.

Discussion of damage due to dust and slags of poor conductivity on bricks of checkerwork regenerators, and how injurious effect can be calculated; effect of dust deposit is made evident by increase of average difference of temperature of brick surface in heating and coal period, and consequent drop of heat-transmission coefficient.

TEMPERATURE MEASUREMENT. Measurement of Gas Temperatures up to 1500 Deg. in Radiation Fields of Variable Antisotropy (Die Messung von Gastemperaturen bis 1500 Grad in Strahlungsfeldern wechselnder Anisotrophie), H. Schmidt. *Archiv fuer das Eisenhuettenwesen (Duesseldorf)*, vol. 2, no. 5, Nov. 1928, pp. 293-298 and (discussion) 298-299, 13 figs.

Functions of gas-temperature measurement; expansion of measuring range up to 1500 deg. cent.; measurement by radiation of thermocouple through hot walls; improvement in design of gas pyrometer; measurements of gas and air temperature in chambers of open-hearth furnace; optical measurement and recording.

OPEN-HEARTH FURNACES—CONTROL

Open Hearth Combustion Process is Regulated Mechanically, G. A. Merkt. *Iron Trade Rev.*, vol. 84, no. 14, Apr. 4, 1929, pp. 914-915, 1 fig.

Details of Isley furnace-control apparatus applied to open-hearth furnace; venturi tubes equipped with motor-driven fans eject furnace gases and supply draft in definite proportions; duplex arrangement of venturi tubes; no moving parts exposed to high temperature.

Investigation of Economy of Different Decarburizing Media for Open-Hearth-Furnace Practice (Untersuchung ueber die Wirtschaftlichkeit verschiedener Frischmittel fuer das Siemens-Martin-Verfahren), S. Schleicher. *Stahl und Eisen (Duesseldorf)*, vol. 49, no. 14, Apr. 4, 1929, pp. 458-462 and (discussion) 462-464, 4 figs.

Based on result of refining tests, it is shown that oxygen content of refining medium has no influence on velocity of melting down of charge; silicic-acid content of refining medium plays important role.

REFRACTORY MATERIALS. Magnesite or Dolomite for Basic Bottoms, M. Karnaoukhov. *Blast Furnace and Steel Plant*, vol. 17, no. 4, Apr., 1929, pp. 545-547.

For majority of plants use of magnesite in making up and repairing hearth is preferable; constitution of magnesite; changes in character of magnesite in making up bottom; solubility in slag; directions for building magnesite hearth; repair of bottoms; comparison of costs of magnesite and dolomite.

REGENERATORS. Heat Exchange in Regenerators (Waermeaustausch in Regeneratoren), H. Hausen. *V. D. I. Zeit. (Berlin)*, vol. 73, no. 13, Mar. 30, 1929, pp. 431-433, 4 figs.

Mathematical analysis of heat exchange in regenerators and recuperators; derivation of formula of efficiency of regenerators.

PICKLING (METALS)

Pickling Sheets with Sulphuric Acid before Galvanizing, H. Bablik. *Iron and Coal Trades Rev. (Lond.)*, vol. 117, no. 3168, Nov. 16, 1928, pp. 732-733, 2 figs.

Hydrochloric acid was formerly used for pickling in iron and steel industry; its place has been taken by sulphuric acid, which is now much cheaper; removal of scale; consumption of pickling solution; tabular data on tests; details of good practice; advantages of machine pickling.

PIG IRON ANALYSIS

Chemical Composition of Pig Iron as an Indispensable Measure of Quality (Die chemische Zusammensetzung ein unzulaenglicher Masstab der Qualitaet), E. Piwowarsky. *Giesserei (Duesseldorf)*, vol. 16, no. 14, Apr. 5, 1929, pp. 318-321.

Reasons for difference in behavior of pig irons of same chemical composition are discussed in light of past studies and latest research.

PIPE, CAST IRON—CENTRIFUGAL CASTING

Centrifugal Casting (Centrifugation de la fonte), Posenti and Scorza. *Revue de Métallurgie (Paris)*, vol. 26, no. 2, Feb., 1929; pp. 90-92, 5 figs.

Review of developments in centrifugal iron castings; description of authors' system of casting pipe centrifugally, and their experience with process at Pesaro, Italy, where pipe of 200 to 350-mm. diam. and 4-m. length were made; with 12 flasks it is possible to produce 5000 to 7000 kg. of pipe in 8 hr. at very low cost. Abstract of paper read before Assn. Technique de Fonderie.

PLATE MILLS

POWER REQUIREMENT. Power Requirement and Efficiency of Plate and Armor-Plate Mills (Kraftbedarf und Wirkungsgrad von Blech- und Panzerplattenwalzwerken), F. Hammerschmidt and H. Babin. *Archiv fuer das Eisenhuettenwesen (Duesseldorf)*, vol. 2, no. 6, Dec. 1928, pp. 367-370, 3 figs.

Analysis of power consumption of driving motor as determined by previous tests on mills in Witkowitz and English and Swedish tests on plate mills; influence of bearing

friction and other losses on power consumption and efficiency; conclusions as applied to other rolling mills.

RAILS

ALLOY STEEL. Special Hard Wearing Steels for Use in Points and Crossings, P. Marthourey. *Int. Ry. Congress Assn.—Bul. (Brussels)*, vol. 11, no. 2, Mar., 1929, pp. 233-249, 35 figs.

General remarks on phenomenon of wear; review of research to find steel offering great resistance to wear; results of tests of chrome-nickel steel for frogs and switches. From Industrie des Voies Ferrées et des Transports Automobiles, Mar., 1928.

MANGANESE STEEL. Medium-Manganese Rail a Success, E. E. Thum. *Iron Age*, vol. 123, no. 14, Apr. 4, 1929, pp. 940-942, 3 figs.

Origin and advantages found by service records of manganese steel rails are discussed; 230,000 tons now in track; records show low rate of wear, head breakage and transverse fissures; no difficulties experienced at mills.

PROPERTIES. Properties of Basic Rail Steel (Die Eigenschaften von Thomas-Schienenstahl), E. H. Schulz and A. Wimmer. *Stahl und Eisen (Duesseldorf)*, vol. 49, no. 12, Mar. 21, 1929, pp. 385-388.

Investigation with view to retention of basic steels as materials for rails; experimental results covering metallurgical properties, strength, and durability; it is claimed that this steel shows no superiority over open-hearth steel.

RAILS, STEEL

HEAT TREATMENT. Heat Treatment of Steel Rails, Marcotte and Martineau. *Fuels and Furnaces*, vol. 6, no. 8, Aug. 1928, p. 1102.

Developments in heat treatment of steel rails by Sandberg process are described. Translated abstract from Revue de Métallurgie. See reference to original article in Eng. Index. 1927, p. 790.

RAILROAD CONSTRUCTION, IRON AND STEEL

Report on Iron-Steel Structures. *Ry. Age*, vol. 86, no. 9D, Mar. 8, 1929, pp. D166-D168 and (discussion) D168-D169.

Study of electric welding of connections, copper-bearing steel; specifications for steel highway bridges; effect of dead load on impact from moving loads on bridges; rolling tests of plates. Report of Am. Ry. Eng. Assn.

REFRACTORY MATERIALS

Refractories: Their Manufacure and Applications, J. F. Hyslop. *World Power (Lond.)*, vol. 10, no. 56, Aug. 1928, pp. 137-144, 7 figs.

Author surveys characteristics of chief refractory materials, silica and silica-alumina, and illustrates main factors which determine their service behavior and use.

MANUFACTURE. Maker and User of Refractories Gain Through Technical Control, L. J. Trostel. *Chem. and Met. Eng.*, vol. 36, no. 3, Mar., 1929, pp. 148-150, 6 figs.

Laboratory control methods of General Refractories Company, Baltimore; laboratory is two-story brick structure, 40 x 55 ft.; with deep basement for storage rooms; full equipped process laboratory on first floor and chemical laboratory on second floor; processes of laboratory control.

PROPERTIES. Steel-Making Refractories, E. R. Walker. *Chem. Eng. and Min. Rev. (Melbourne)*, vol. 21, no. 244, Jan. 5, 1929, pp. 133-147.

Impurities and their effect on properties of fireclay refractories; desirable properties for various types of bricks and other forms are discussed; silica bricks; siliceous sand; ganister; basic refractories, as magnesite and dolomite; neutral refractories. (Continuation of serial.) Read before Bureau of Steel Mfrs. of Australia.

TESTING. A General View of the Testing of Refractories, J. S. McDowell. *Am. Refractories Inst.—Tech. Bul.*, no. 28, Oct., 1928, 8 pp.

Description of following tests: pyrometric cone equivalent; load, spalling, reheating, slag, and control tests.

Present Status of Tests for Refractories, S. M. Phelps and J. S. McDowell. *Am. Soc. Mech. Engrs.—Advance paper for mtg.* May 13 to 16, 1929, 4 pp., 5 figs.

Tests for refractories divided into three general types; durability, suitability, and control; pyrometric cone equivalent; standard load test; reheat test; water-immersion spalling test; use of tests in selection of refractories.

ROLLING MILLS

ANTI-FRICTION BEARINGS. Developments in the Application of Anti-Friction Bearings to Roll Necks, E. C. Gainsborg. *Rolling Mill Jl.*, vol. 3, no. 1, Jan. 1929, pp. 19-22.

Factors determining useful service to be obtained from anti-friction bearings; measurement of roll-neck pressure; various types of bearings available; method of lubrication.

ANTI-FRICTION BEARINGS. Selection of Anti-Friction Bearings for Roll Necks, J. H. Van Campen. *Rolling Mill Jl.*, vol. 3, no. 1, Jan. 1, 1929, pp. 31-34, 4 figs.

Conditions governing successful application of roller bearings to roll necks; types of anti-friction bearings, and typical installations.

ELECTRIC EQUIPMENT. Electrical Developments in the Iron and Steel Industry, D. W. Dean. *Rolling Mill Jl.*, vol. 3, no. 1, Jan. 1929, pp. 23-26 and 34, 4 figs.

Recent applications of adjustable and constant-speed drives to rolling mills; welded electric machinery; steel-mill auxiliaries; electrification of present mills.

FOUR-HIGH. Power Efficiency in Rolling a Function of the Roll Diameter, L. Weiss. *Rolling Mill Jl.*, vol. 2, no. 10, Oct. 1928, pp. 401-402.

It is shown that with four-high mills, operating economy is to some extent dissipated by elastic compression of working and supporting mills; actual saving effected is explained as due to fact that drawing process takes place on portion of mill train under very favorable conditions. Abstract translated from Zeit. fuer Metallkunde, Aug. 1928.

RAIL. Mill to Roll Split Rails into Bars. *Iron Age*, vol. 123, no. 3, Jan. 17, 1929, pp. 208-210, 3 figs.

Methods in new 10-in. mill erected by Barton Spiderweb System, Chicago; mill con-

sists of six three-high 10-in. stands, four designated as break down units; rerolling rails will be hot split midway between tread and flange in first break-down stand; rolls in first and succeeding passes so designed that head and flange sections of rail will be rolled concurrently with one heating.

ROLL HEATERS. Electric-Induction Roll Heaters. *Foundry Trade Jl.* (*Lond.*), vol. 39, no. 644, Dec. 20, 1928, pp. 453-454, 7 figs.

In order to ensure that rolls of tinplate shall be cylindrical during normal rolling, new method has been devised; principle of this method of roll heating consists of surrounding pair of rolls with coil of wire through which flows alternating current; this induces currents in mass of metal contained in rolls, so generating heat therein; thus difficulty of rolls being hotter at center of their surface is counteracted.

ROLLING MILL PRACTICE

Better Control Typifies Furnace and Mill Progress in 1928. *Iron Trade Rev.*, vol. 84, no. 1, Jan. 3, 1929, pp. 21-22, 2 figs.

Refinement of present-type open-hearth furnace; larger furnaces operated; increasing desire to conserve fuel; special checkers employed; important developments in steel rolling made in secondary or finishing mills rather than in extension of primary capacity; radical changes made in practice of producing wide thin sheet for strip material; synchronous motors favored; opinions of leading authorities quoted.

Roll Pass Design, W. Trinks. *Rolling Mill Jl.*, vol. 3, no. 1, Jan. 1929, pp. 13-18, 5 figs.

General introduction to roll-pass design; definition of terms, description of various types of rolling mills, passes, rolls; problems of design. (To be continued.)

METAL FLOW. Experimental Investigations on Metal Flow in Rolling, N. Metz. *Rolling Mill Jl.*, vol. 2, no. 10, Oct. 1928, pp. 419-422, 11 figs.

Article is based on information obtained by experiments at German Iron and Steel Institute on flow of metal in various shapes of passes; material flow in square pass formed from oval, in round produced from oval, in rail upsetting passes.

RUST PREVENTION—RESEARCH

Lanoline Rust Prevention, C. Jakeman. *Department of Sci. and Indus. Research—Report (Lond.)*, no. 12, 1929, 22 pp., 8 figs., partly on supp. plates.

Citation of report in earlier investigation, as published in Engineering, vol. 120, 1925, pp. 123-125; coatings of lanoline grease appeared to be superior to petroleum grease; further tests are now reported; sources of lanoline; solvents for lanoline; colored solutions, to ensure covering all parts; wax-hardened mixtures; quality of lanoline coating has not yet been determined; method of application, and cost of materials; specifications for materials.

SAWS, METAL WORKING

A Bench Metal Saw, H. Haeneke. *Eng. Progress (Berlin)*, vol. 10, no. 4, Apr., 1929, pp. 107-108, 2 figs.

Small high-speed circular saw is intended to replace hack saws and other hand saws in cutting bars of small cross-section; with saw diam. of 4 in. this tool cuts at one setting round stock of diam. of 1.2 in., channel iron up to 1.2 in. by 1.3 in. and corresponding sizes of other shapes.

SHEET MILLS

Reversing Sheet Mill Development to Eliminate Heat Losses, L. Cammen. *Iron Trade Rev.*, vol. 85, no. 15, Apr. 11, 1929, pp. 994-995, 2 figs.

Details of reversing 2-high mill, which, with crew of three men, will roll in excess of 2 tons of sheet steel per hour; mill is now under construction in United States; roughed sheets are supplied to left-hand end of coil which is high-frequency reheating coil; screwdown operated electrically; mill is well adapted to small-mill operation with limited output.

Sheet Mill Furnace Installation at Youngstown Sheet and Tube Company, F. W. Manker. *Fuels and Furnaces*, vol. 7, no. 1, Jan. 1929, pp. 109-111, 4 figs.

Sheets are annealed in continuous and automatic kiln-type furnaces, pickled and then tinned in 20 automatic tinning machines, each tinning unit consisting of tin pot, tinning machine, and combustion chamber with system by which close control of temperature is obtained.

SHEET MILL PRACTICE

Rolling High Grade Auto Sheets, W. H. Melaney. *Blast Furnace and Steel Plant*, vol. 17, no. 1, Jan. 1929, pp. 91-93, 1 fig.

Double-mill loose-pack system produces better surface, it is claimed; proper type of rolls for roughing mills; making rolls for sheet mills; turning rolls for double-mill system; steam on rolls; increasing roll diameters.

Developments in Sheet Rolling Practice, L. Cammen. *Rolling Mill Jl.*, vol. 2, no. 12, Dec. 1928, pp. 513-516.

Distinctions between sheet rolling and strip rolling, and between hot and cold rolling; conventional rolling practice; influence of labor on cost and temperature of rolling; wide strip rolling; heating between passes; American vs. German practice. Abstract of paper presented before Am. Soc. Mech. Engrs.

Steel-Sheet Rolling in the United States, L. Cammen. *Iron and Coal Trades Rev.* (*Lond.*), vol. 117, no. 3173, Dec. 21, 1928, p. 910.

Author criticizes existing practice in United States and elsewhere; wide-strip mill regarded as more or less a speculation; finishing of sheet constitutes major part of cost; economic salvation of sheet industry lies in reorganization of entire method of production; principles enumerated, involving approach to complete mechanization. Brief abstract of paper before Am. Soc. Mech. Engrs., including discussion.

SHEET STEEL

AUTOMOBILE. Characteristics of Auto-Body Sheets, R. C. Todd. *Iron Age*, vol. 123, no. 15, Apr. 11, 1929, pp. 1006-1008.

Present standards regarding automobile sheet grades are discussed; developments of past decade which have resulted in definite standards for various grades; properties demanded for more economical automobile manufacture; auto-body grade; auto-body deep and extra deep drawing; hood and fender deep and extra deep drawing; radiator casing steel.

HANDLING. The Baker Sheet Handler. *Iron and Coal Trades Rev. (Lond.)*, vol. 118, no. 3185, Mar. 15, 1929, pp. 398-399, 3 figs.

Brief discussion of unsatisfactory and troublesome methods of handling steel sheets in shipments; description of Baker sheet handler, developed by Baker-Raulang Co. of Cleveland, Ohio, will load about 2000 tons of sheets in 10-hr. day with 1 operator and 1 helper; 2 men can load about 80 tons per day by hand.

RESEARCH. Performance of Special Steel Sheets. *Heat Treating and Forging*, vol. 15, no. 3, Mar., 1929, pp. 328-329.

Study of results attained by use of recently developed drawing sheets; savings in cost as well as dependability in production cited; production methods for making automobile radiators; reduced breakage; reduced grinding labor.

SPONGE IRON—MANUFACTURE

Sponge Iron by the Smith Process, G. B. Waterhouse. *Iron Age*, vol. 123, no. 17, Apr. 25, 1929, pp. 1143-1145, 4 figs.

Description of process of General Reduction Corp., Detroit, for production of fine sponge iron; vertical ovens like coke ovens used; 100 per cent reduction claimed; process will directly assist blast furnace.

SPRINGS

Springs, P. Gradenigo. *Engineering (Lond.)*, vol. 126, no. 3267, Aug. 24, 1928, pp. 240-241.

Review of Italian book published by S. Lattes & Co. in Turin; author makes exhaustive study of springs of all types; divided into four volumes, of which first deals with properties of materials suitable for manufacture of springs; metallurgy of spring steels; second volume is mainly devoted to manufacturing methods; third deals with testing both raw material and finished springs; in last volume various methods of calculating characteristics of different types of spring are given.

SPRINGS—MANUFACTURE

Springs for Service at High Temperatures, J. R. Fritze, R. J. Sutton and F. R. Porter. *Heat Treating and Forging*, vol. 15, no. 3, Mar., 1929, pp. 301-302 and 310, 6 figs.

Selection of steel of suitable composition and proper treatments to facilitate its fabrication into springs is discussed; manufacturing operations described; elimination tests for steel; annealing and tempering; testing.

SPRING STEEL

TESTING. Researches on Springs—Torsional Fatigue Tests on Spring Steels, G. A. Hankins. *Dept of Sci. and Indus. Research—Eng. Research (Lond.), Special Report* no. 9, 1928, 24 pp., 8 figs. partly on supp. plates.

Investigation by National Physical Laboratory as part of systematic study undertaken by Springs Research Committee; tests devised to disclose actual properties under torsional stress cycles similar to those under working conditions; variations due to surface blemish, decarburization and other disturbing factors eliminated; results given in full with discussion.

STEEL

AIRCRAFT—FAILURES. Steel Failures in Aircraft, F. T. Sisco. *Am. Soc. Steel Treating—Trans.*, vol. 15, no. 4, Apr., 1929, pp. 589-624 and (discussion) 624-629, 38 figs.

Failures in steel aircraft parts investigated by Metallurgical Laboratory at Wright Field are described; included in this classification is account of failures due to faulty manufacturing methods and not caused by defective material; failures due to internal defects such as non-metallic inclusions, and due to seams and other surface defects; miscellaneous failures due to defective steel; failures due to faulty or wrong heat treatment, and to welding.

COLD WORKING. Effects of Cold Drawing on the Strength and Endurance of Mild Steel, R. M. Brown. *Rolling Mill Jl.*, vol. 3, no. 3, Mar., 1929, pp. 123-126.

Relationship between strain and energy are considered; effects of fluctuating stresses on tensile strength; creep under pulsating loads. (Continuation of serial.)

GAS CONTENT. Injurious and Beneficial Effects of Gases in Steel (Ueber schaedliche und nuetzliche Wirkungen der Gase im Stahl), F. Rapatz. *Zeit. fuer Metallkunde (Berlin)*, vol. 21, no. 3, Mar., 1929, pp. 89-92 and (discussion) 92-93, 8 figs.

Oxygen is cause of soft spots occurring in connection with hardening; nitrogen, in quantities present in steel, is harmless; in case-hardening it is used to advantage, either alone or combined with carbon; gas formation with solidification promotes formation of small crystallites.

MAGNET. Cobalt Chrome Magnet Steels, H. E. Kershaw. *Indus. Chemist (Lond.)*, vol. 5, no. 49, Feb., 1929, pp. 47-48, 2 figs.

Analysis of air-hardening and oil-hardening cobalt steels and comparison between magnetic values of old tungsten and new cobalt steels; development of graph showing hysteresis loop of magnet.

MAGNETIC PROPERTIES. Magnetic Investigations of Carbon Steel, C. C. Duell. *Am. Soc. Steel Treating—Trans.*, vol. 15, no. 4, Apr., 1929, pp. 630-638 and (discussion) 638-651, 20 figs.

Results obtained by magnetic methods in investigation of iron and steels of various carbon contents and in various conditions; in steels containing cementite, drop in magnetism occurs from 150 to 220 deg. cent.; in steels containing only martensite decreased magnetism occurs at 300 deg. cent.; cementite is stated to be stable compound at all temperatures below 745 deg. cent., and with proper carburizing heat treatment below this temperature, iron may be converted entirely into carbide.

PROTECTIVE COATINGS—CORROSION. The Electrochemical Corrosion of Painted or Lacquered Steel, U. R. Evans. *Am. Electrochem. Soc.—Advance paper for mtg.* May 27, to 30, 1929, pp. 37-58, 2 figs.

Observations regarding action of different pigments in arresting or stimulating corrosion of painted or lacquered steel; conclusions reached are in general accord with those of Cushman and Gardner in America and of Ragg in Germany.

STAINLESS STEEL

Heat and Corrosion Resistant Alloys, T. H. Nelson. *Iron Age*, vol. 123, no. 17, Apr. 25, 1929, pp. 1139-1142, 5 figs.

Description of stainless steel, what it is and how it was discovered; properties of three chief types; use of chrome irons for structure.

STEAM POWER PLANTS

METALS. New Metals Now Used for Modern High Pressures and Temperatures, A. E. White. *Power*, vol. 69, no. 8, Feb. 19, 1929, pp. 330-332.

Power plants are now in state of transition to use of special metals and alloys such as experienced by machine and automotive field some years ago; use of cast iron decreasing rapidly; how alloy steels are used; stainless steel used for turbine blades; use of welding in power plant construction is receiving increased attention; magnetic testing for detecting flaws in metals. Abstract of paper presented before Midwest Power Eng. Conference.

STEEL ANALYSIS, X-RAY

X-Ray Study of Structure of Hardened Carbon Steel (Eine roentgenographische Untersuchung der Struktur des gehaerteten Kohlenstoffstahls), G. Kurdjumow and E. Kaminsky. *Zeit. fuer Physik (Berlin)*, vol. 53, no. 9 to 10, 1929, pp. 696-707, 5 figs.

Results of experimental study at Leningrad Laboratory of Engineering Physics, showing presence of tetragonal structure within body, not only at surface, of specimens; effect of annealing at 100 deg. cent.

X-Ray Investigation of the Internal Stress of Carbon Steels, S. Sekito. *Tohoku Imperial Univ.—Science Report (Tokyo)*, vol. 17, no. 7, Dec., 1928, pp. 1227-1236, 5 figs.

Variation of lattice constant and internal stress in cold-worked carbon steels; internal stress found little less than tensile strength; this confirms Honda's view that internal stress due to cold working cannot exceed tensile strength. (In English.)

STEEL CASTINGS

Steel Castings from Moulder's Viewpoint, J. Jefferson. *Foundry Trade Jl. (Lond.)*, vol. 39, no. 641, Nov. 29, 1928, pp. 393-394, 3 figs.

Simple tests were carried out and following summary is given: designed dimensions must be adhered to; material must be everywhere homogeneous; it must be free from internal cavities, from cracks, and from porosity; internal stresses must be reasonably absent; founder cannot do more than attempt to correct by annealing; reduction of thickness. Paper presented before joint meeting of Inst. Brit. Foundrymen and Inst. Mech Engrs.

STRUCTURE. Abnormalities in the Structure of Steel Castings (Ueber anormale Erscheinungsformen im Gefuege von Stahlguss), E. Piwowarsky. *Stahl und Eisen (Duesseldorf)*, vol. 48, no. 48, Nov. 29, 1928, pp. 1665-1669, 15 figs.

Tendency to network formation of pearlite after change in crystallization in connection with deoxidation with aluminum and vanadium; there is as yet no satisfactory explanation of this phenomenon; number of photomicrographs are shown.

STEEL CASTINGS—GRINDING

Efficient Steel Casting Grinding, F. A. Lorenz, Jr. *Am. Foundrymen's Assn.—Preprint for mtg.* Apr. 8-11, 1929, pp. 199-204, 1 fig.

By just giving grinder efficient service and bringing work to him in properly designed container, giving him continuous work and service enabling him to dispose of ground castings, efficiency of unit has been increased 30 per cent; grinding wheel only one of several factors; question of layout should be considered; some actual results; one operator now replaces five operators.

STEEL CONSTITUENTS

Effect of the Principal Additional Elements in Steels, L. Persoz. *Foundry Trade Jl. (Lond.)*, vol. 40, no. 655, Mar. 7, 1929, pp. 181-182.

Subject is discussed in relation to following: nitrogen, phosphorus, arsenic, sulphur, manganese, silicon, chromium, nickel, copper, tungsten, vanadium, molybdenum, uranium, cobalt, titanium; aluminum, boron, and cerium. Translated from Aciers Spéciaux et leurs Emplois, Nov., 1928.

STEEL CORROSION

The Corrosion of Steel by Acid Solutions, H. Endo. *Tohoku Imperial Univ.—Science Report (Tokyo)*, vol. 17, no. 7, Dec., 1928, pp. 1245-1263, 15 figs.

Results of experiments on corrosion of steels when immersed in hydrochloric, sulphuric, or nitric-acid solutions, steels containing various amounts of carbon, cobalt, nickel, chromium, manganese, molybdenum, vanadium, etc.; relation between dissolution of hardened steels in sulphuric-acid solution and degree of tempering has also been studied. (In English.)

STEEL DEOXIDATION

Deoxidation of Steel with Silicon, C. H. Herty, Jr. and G. R. Fitterer. *Am. Soc. Steel Treating—Trans.*, vol. 15, no. 4, Apr., 1929, pp. 569-584 and (discussion) 584-588, 8 figs.

Outline of function of silicon in basic open-hearth process; formation and elimination of silicates; mechanism of deoxidation with silicon is: (1) formation of silica by reversible reaction of silicon with ferrous oxide, and (2) fluxing of more ferrous oxide by silica, turning ferrous-silicate inclusion; complete deoxidation of steel is impossible unless oxide formed is absolutely infusible and insoluble at steel-making temperatures.

STEEL FOUNDRIES

OPERATION AND SALES COOPERATION. Co-operation Between Operating and Sales Departments Is Essential for Success, K. V. Wheeler. *Foundry*, vol. 56, no. 23, Dec. 1, 1928, pp. 968-970.

Why, in steel foundry, it is so essential that operating and sales departments should always work together closely, is discussed; making of any casting in steel is not solely producing problem, but combination of producing and engineering efforts, which vary with type, size and design; design of castings; planning for customer's requirements. Abstract of paper presented before Am. Foundrymen's Assn.

News of the Society

NATIONAL METAL CONGRESS AND EXPOSITION

INTERNATIONAL attention is being given the first National Metal Congress to be held in Cleveland, September 9 to 13, simultaneously with the National Metal Exposition, which will be housed in the spacious Cleveland Public Auditorium.

At the exposition will be displayed the latest machinery produced for the production, treating, fabrication, welding and use of metals. Already more than 90 per cent of the space available for exhibits has been subscribed for by more than 200 exhibitors. It is estimated that more than 75,000 people will see the products and machinery which will be displayed at the exposition.

The National Metal Congress means the assemblage in Cleveland of the leading minds of the field of metals for an exchange of ideas and for the promotion of further advancement in the great industries in which every metal is developed and consumed.

More than 150 experts in the field of metals will deliver papers at the National Metal Congress. These papers will include a complete detailed discussion of the various and important developments in the metal industry of the past year, some of them revolutionary.

The importance of the papers and discussions will bring to the Congress the greatest assembly of metal scientists that this country has had together.

Sponsoring this Congress and the National Metal Exposition are the American Society for Steel Treating, the American Welding Society, the Institute of Metals Division and the Iron and Steel Division of the American Institute of Mining and Metallurgical Engineers and the Iron and Steel Division of the American Society of Mechanical Engineers. Hotel headquarters for these groups will be the Hotel Cleveland, Hotel Statler, Hotel Cleveland and Hollenden Hotel respectively.

These societies represent a combined membership of more than 15,000 men in the metal industries and a great majority of these are arranging to attend the Congress and Exposition.

On the following pages there appears the list of papers which will be presented during the week by the five societies.

Technical Papers Program of the American Society for Steel Treating

DILATATION OF STEEL DURING QUENCHING—G. M. Eaton, Molybdenum Corporation of America, Pittsburgh.

ON THE DISTRIBUTION OF HARDNESS PRODUCED BY COLD WORKING—W. P. Sykes, Wire Division, General Electric Company, and A. C. Ellsworth, Thompson Products Company, Cleveland.

THE ECONOMICAL REUSE OF SOLID CARBURIZING MATERIALS—H. B. Knowlton, The International Harvester Company, Fort Wayne, Ind.

THE SELECTION OF CASE HARDENING STEELS FOR HIGHLY STRESSED GEARS—H. W. McQuaid and O. W. McMullan, Timken-Detroit Axle Co., Detroit.

ugust

l Con-
th the
veland

or the
more
for by
le will
on.
of the
or the
metal

at the
ed dis-
stry of

ongress
gether.
are the
the In-
nerican
eel Di-
quarters
eveland

00 men
attend

will be

ting

um Cor-

—W. P.
lsworth,

nowlton,

GEARS—
Detroit.